Transactions of
American Society for Steel Treating

Vol. XI May, 1927 No. 5

FEATURES

Hardness Testing of Steel Balls by Magnetic Methods—
S. R. Williams 677

Graphitizing Behaviour of Iron Carbide in Pure Iron-Carbon Alloys in the Critical Range—
H. P. Evans and Anson Hayes 691

High Speed Steels— *F. C. A. H. Lantsberry* 711

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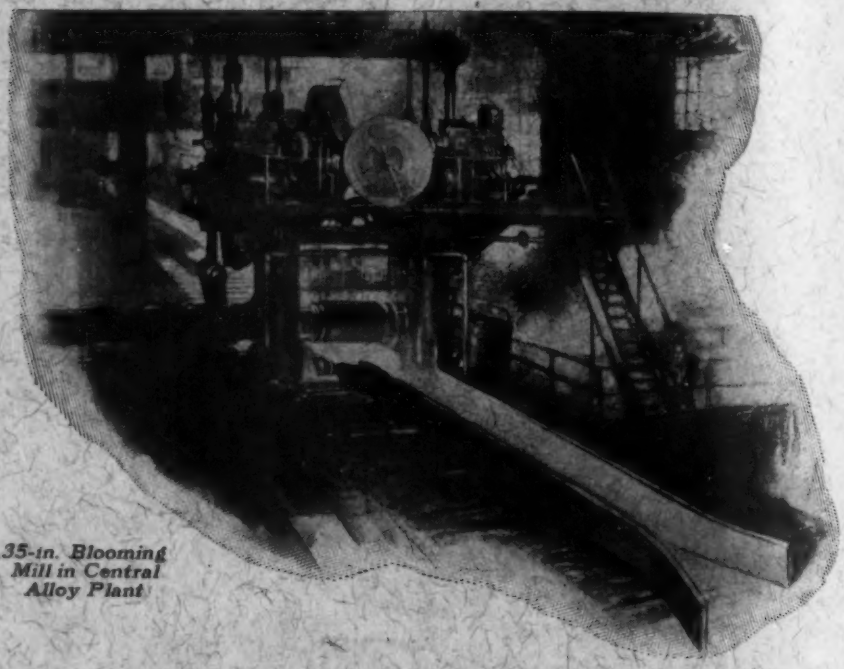
Facts and Principles Concerning Steel and Heat Treatment—Part XII— *H. B. Knowlton* 791



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THE

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RAY T. BAYLESS, *Editor*

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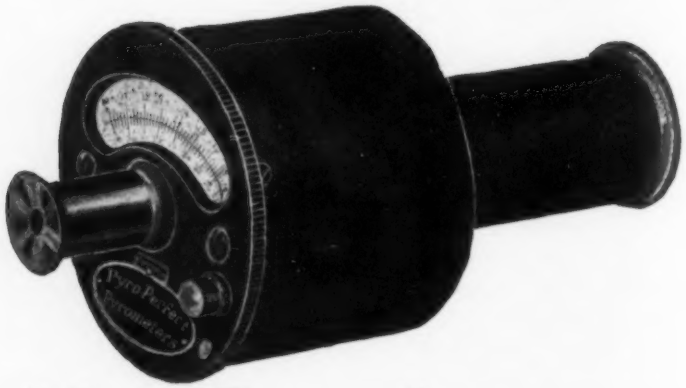

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TRANSACTIONS
American Society for Steel Treating

VOL. XI MAY, 1927 NO. 5

THE SPRING SECTIONAL MEETING

COMPLETE arrangements have been made for the entire program of the Spring Sectional meeting to be held in Milwaukee, May 19 and 20, 1927. The energetic Milwaukee chapter committees have done a fine piece of work in preparing the program of events for this 2-day session. The Hotel Pfister will be headquarters.

On Wednesday, May 18, the national officers and directors will hold an all-day meeting at the Hotel Pfister, to transact important matters of business relative to the society. Also the Recommended Practice Committee of the Society will hold a meeting in the headquarters hotel the same day. In the evening the usual Early Birds' Dinner and theater party will be held.

Thursday and Friday, May 19 and 20, will be devoted to technical sessions and plant inspection trips. Papers will be presented by men who are authorities in their lines, and we are sure that their papers will carry many new ideas and much food for thought. The authors who will present papers are Messrs. T. McLean Jasper, Harold F. Wood, R. S. Archer, R. M. Sandberg, H. B. Northrup and J. B. Green. The subjects of the papers to be presented will be found in the scheduled program on page 676.

The program for the banquet Thursday evening will be a real treat. The society is to be honored by having as its principal speakers S. M. Havens, general manager of the Wyman-Gordon Co., Harvey, Ill., and Dr. Max Mason, president of the University of Chicago. The subject of these speeches is given on the next page.

In addition to the plants scheduled for inspection Thursday and Friday afternoon, the committee in charge has made arrangements so that members and guests will be able to inspect the following plants: Crucible Steel Co., Sivyer Steel Castings Co., Maynard Steel Casting Co., Geo. H. Smith Co., Milwaukee Steel Casting Co., Harley Davidson Co., Ladish Drop Forge Co., Falk Corporation, Nordberg Mfg. Co., Bucyrus Co., Harnischfeger Corporation, Nash Motors, Milwaukee Plant, Globe Seamless Tube Co.

PROGRAM FOR SPRING SECTIONAL MEETING, HOTEL PFISTER, MILWAUKEE, MAY 19-20, 1927

WEDNESDAY, MAY 18

10:00 A. M.—Meeting of Officers and Board of Directors.
10:00 A. M.—Meeting of Recommended Practice Committee.
6:00 P. M.—Early Birds' Dinner and Registration—Hotel Pfister.

THURSDAY, MAY 19

9:30-10:00 A. M.—Registration. Ball Room, Hotel Pfister.

Morning Technical Session Dr. J. A. Mathews, Chairman

10:00-11:00 A. M.—Scientific Research and Its Application to Industry—T. McLean Jasper, Director of Research, A. O. Smith Corp., Milwaukee.

11:00-12:00 A. M.—The Automatic Electric Furnace—Its Uses and Possibilities—Harold F. Wood, Chief Metallurgist, Ingalls Shepard Division, Wyman-Gordon Co., Harvey, Illinois.

Afternoon Session

1:00- 4:30 P. M.—Luncheon and plant inspection at the plants of the Allis Chalmers Manufacturing Co.

Evening Session

6:30 P. M.—Banquet and Entertainment —Ball Room, Hotel Pfister.

Speakers

S. M. Havens, General Manager, Wyman-Gordon Co.
The Technical Organization in Industry
Dr. Max Mason, President of Chicago University
Formalism and the Deeper Insight

FRIDAY, MAY 20

Morning Technical Session Prof. H. M. Boylston, Chairman

10:00-11:00 A M.—Heat Treatment of Wrought Aluminum Alloys of High Strength, R. S. Archer, Research Bureau, Aluminum Company of America, Cleveland.

11:00-12:00 A. M.—Application of Theory to Practice in Heat Treatment of High Speed Steel—R. M. Sandberg, Assistant Manager, Columbia Tool Steel Co., Chicago Heights, Ill.

Afternoon Session

1:00- 5:00 P. M.—Plant Inspection—Optional Selection of either A. O. Smith Corporation or International Harvester Co.

Evening Technical Session Dr. F. C. Langenberg, Chairman

7:00- 8:00 P. M.—Case Carburization of Steels by Means of Salt Baths of Low Cyanide Concentration—H. B. Northrup, J. W. Kelley Co., Cleveland.

8:00 P. M. —Processing Welding Wire and Its Effect on the Flow of Welding Metal, J. B. Green, Research Department, Chicago Steel and Wire Co., Chicago (Illustrated with Stereopticon and Movies).

HARDNESS TESTING OF STEEL BALLS BY MAGNETIC METHODS

By S. R. Williams

Abstract

In a former paper,[1] published in these Transactions, *a magnetic method for testing the hardness of steel balls was described. The present paper deals with improvements in that method and a better technique in manipulation as now worked out for the outfit.*

Introduction

THE problem of testing steel balls before they are put into actual service seems to be one of considerable interest if one may judge by the reactions which come in reply to letters sent out to steel ball manufacturers. The correspondence shows also that if a practical method of testing steel balls magnetically is to be produced there is a need for co-operation between those interested in the field of pure science in magnetism and those who are interested in finding out what tests will indicate when a first class steel ball is produced.

Of all the methods now used for testing the hardness of steel balls the magnetic method seems to offer the greatest promise because no injury is offered to the ball which is being tested. Along with the hardness tests must also go wearing qualities and resistance to crushing loads. Eventually it will be possible to find definite relations between magnetic and all other physical properties of bodies for there seems to be a fundamental principle that there is one and only one set of physical and chemical properties corresponding to a given set of magnetic characteristics, and conversely, there is one and only one set of magnetic properties corresponding to a given set of physical and chemical characteristics. This offers a large field of research concerning the relations between magnetic properties and the mechanical hardness of substances.

[1] S. R. Williams, Transactions, American Society for Steel Treating, Vol. 5, 1924, p. 479.

A paper presented before the eighth annual convention of the Society, Chicago, September 20 to 24, 1926. The author Dr. S. R. Williams is professor of physics, Amherst College, Amherst, Mass.

Apparatus and Manipulation

The method of testing steel balls, magnetically, rests upon the principle that the external fields of permanent horse-shoe magnets or of electromagnets are distorted when a steel ball is placed between their poles. The greatest distortion will occur

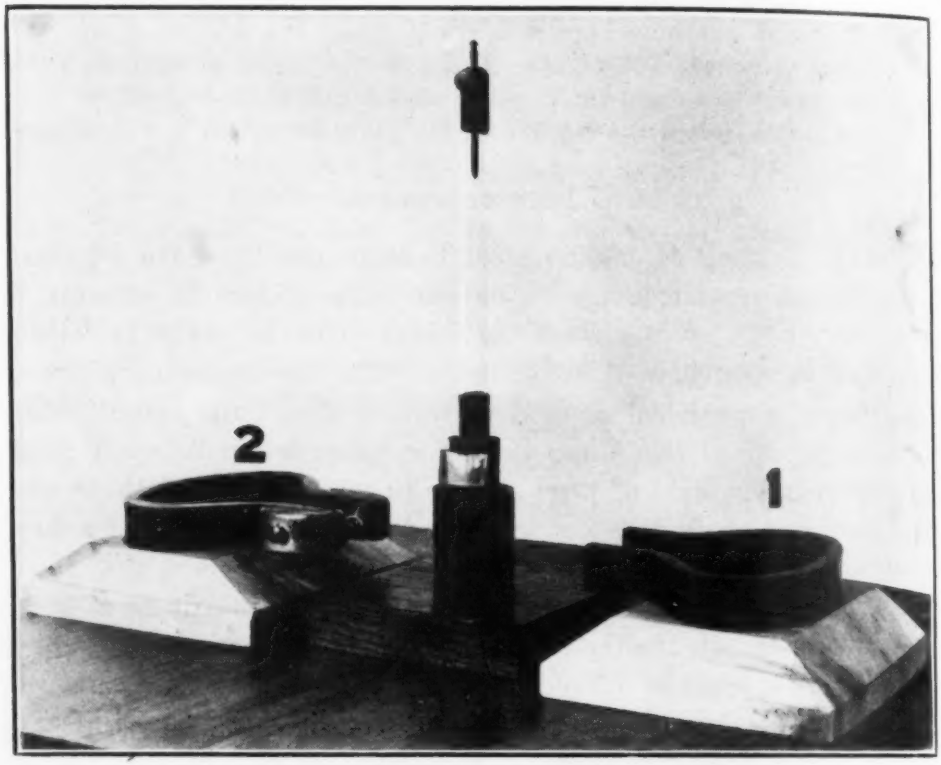


Fig. 1—Photograph Showing the Set-Up of the Magnetic Hardometer.

when the permeability is largest, which in general is true the softer the ball used. If a magnetometer is placed in this field, then the position of the needle in the magnetometer will be different when the ball is present than when it is removed. A soft ball should give a larger deflection of the magnetometer than a hard one.

In the former method the external field in which the magnetometer was placed was the resultant of the earth's field and that of a single horse-shoe magnet. This meant that as the earth's field fluctuated the deflections of the magnetometer varied, and the more so the farther the magnetometer was placed from the magnet.

In the present arrangement of magnetic fields the magnetometer is placed between two horse-shoe magnets, whose fields oppose each other as shown in Fig. 1. In this "set-up" the field in which the magnetometer is placed is affected measurably only by the variations produced by the introduction or withdrawal of the steel ball between the poles of the magnet. The distance

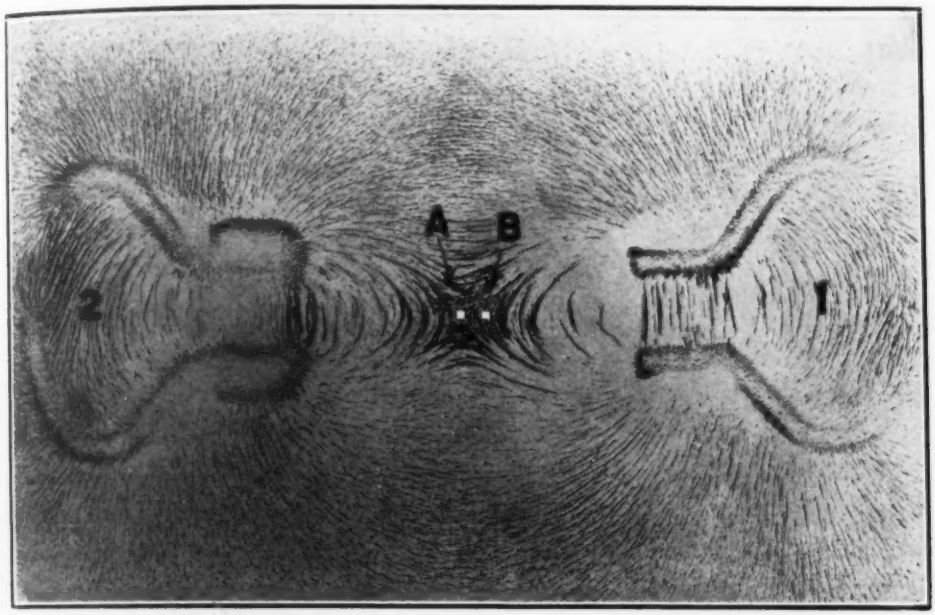


Fig. 2—The Field Between the Two Horseshoe Magnets Shown in Fig. 1. A is the Position of the Neutral Point When the Ball is in Position and B When the Ball is Removed.

apart of the pole ends of the two magnets is about 20 centimeters (7.8 inches) which brings the neutral point about 10 centimeters (3.9 inches) from each magnet. In Fig. 2 is shown the resultant field of the two magnets when the steel ball is between the poles of the magnet No. 2. The neutral point of the field is indicated by the point marked A. The point B marks the neutral point of the field when the ball is removed, thus showing that the introduction of the steel ball shifts the neutral point about half an inch. In this particular case the ball was an ordinary steel ball $\frac{7}{16}$ of an inch in diameter. Fig. 2 also shows how sharply the neutral point is defined so that if a short magnetic needle is used the displacement of the neutral point by the steel ball would cause a deflection of the needle through 180 degrees. A long needle, like those used in surveyor's transits, would indicate very little change in the field. The magnet used for a magnetometer in

this work was bell-shaped, and had a length vertically of about 2 centimeters, (0.75 inch), the diameter being about 1 centimeter. It was made of tungsten steel because Miss Moir[2] has shown that tungsten steel gives greater intensities of magnetization after aging than do other kinds of steel. This bell-shaped magnet was fastened to a stiff brass shank to which the mirror was also attached and the whole system suspended by a phosphor-bronze fiber. In order to dampen the oscillations, the magnet swung in a heavy brass cup, which, in turn, was filled with a light transformer oil.

The deflections of the magnetometer were read by a spot of light reflected from the mirror. Inasmuch as the scale readings became an arbitrary set of magnetic numbers for hardness, the scale was reversed so that deflections giving small numbers corresponded to soft and large numbers to hard balls. Inasmuch as the length of the magnetometer needle has a great deal to do with the sensitivity of the outfit, the position of the magnetometer between the two horse-shoe magnets seems to be a matter of "cut-and-try". Once a good location has been found the whole outfit may be picked up, transported and turned in any direction desired. Standardization of any magnetic hardometer must be in terms of a standard steel ball.

It seemed desirable that when a given sized steel ball was placed between the poles of the magnet, No. 2, Fig. 1, that it should come as near filling the gap as possible so extra pole-pieces were attached to the permanent magnets as shown in Fig. 3. This cut shows the ball in position, being held there by the magnetic attraction of the pole. In the faces of the pole-pieces small conical depressions were made which insured the steel balls being seated each time in exactly the same position. Fig. 1 shows only one of the magnets with the extra pole-pieces. In practice both magnets carried the extensions of poles so that with the two different sized gaps between the poles several sizes of steel balls could be tested without changing the extra pole-pieces.

The horse-shoe magnets were obtained from the Rawson Electrical Instrument Company of Cambridge, Mass. The total flux of these magnets was 17500 maxwells for magnet No. 1 and 17890 for magnet No. 2 as against only 7250 maxwells for the magnet

[2]Moir, *Philosophical Magazine*, 28, 1914, p. 738.

used in the outfit described in the former paper. A criticism of the former device was that it magnetized the steel ball and when they were rotated between the poles most any deflection of the magnetometer could be obtained because of retentivity. To obviate this difficulty the balls were placed directly between the poles without rotation. If a second reading was desired the ball

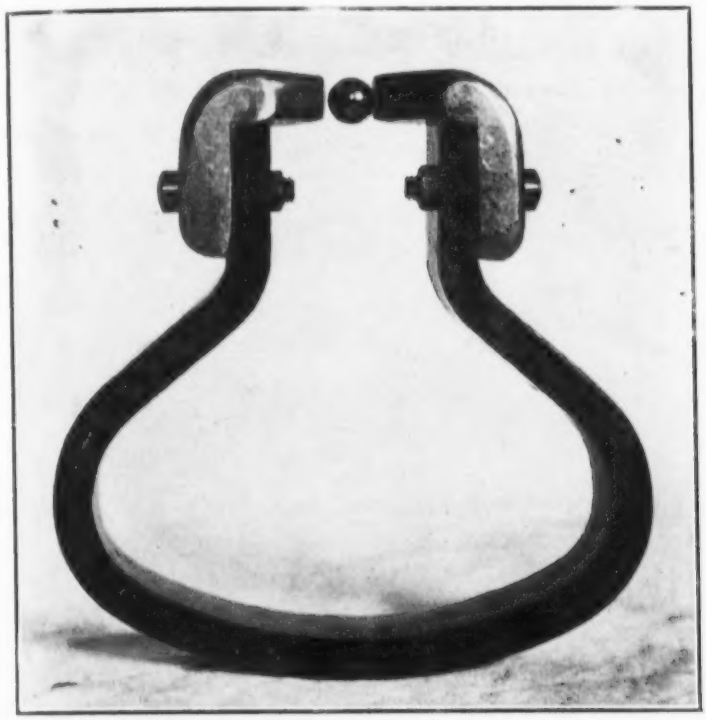


Fig. 3—Photograph Showing the Extension of Poles, and the Steel Ball in Position.

was removed and allowed to roll through a solenoid through which an alternating current was flowing. This coil was 32.9 centimeters long and 3.5 centimeters in free diameter with a constant of 44.8 gauss per ampere. When an alternating current of 110 volts was sent through it a current of about 7.5 amperes flowed in the coil. The ball passing through the strong alternating field at the center of the solenoid and then on through and out some distance from it was found to be thoroughly demagnetized when it could again be placed in the magnetic hardometer and tested as before. Balls classified for different degrees of hardness showed a similar classification after having been demagnetized in the manner described. There was some indication that a few showed

a small increase in hardness. It has been noticed elsewhere that the process of magnetization changes slightly the hardness of some steels.[3]

As a matter of routine testing it would seem best to make just one reading on each ball and let it go at that, for as soon as the ball is placed in the field it becomes permanently magnetized and when rotated the permanent field of the steel ball affects the resultant field in which the magnetometer is placed. If a steel ball is placed between the poles and left there a few seconds and then removed it will show magnetic polarity as in Fig. 4, where

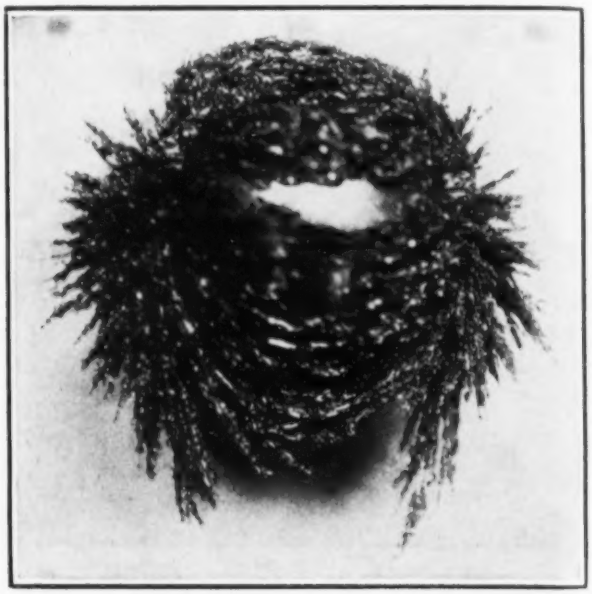


Fig. 4—Photograph of a Permanently Magnetized Steel Ball. The Polarity is Shown by the Arrangement of the Iron Filings.

the ball has been dipped in iron filings. If now the ball is replaced in the field without demagnetization but with the line through the magnetic poles of the ball at right angles to the field in which it is placed it will show four poles when dipped in iron filings. This indicates, of course, why the rotation of the ball gives various readings as the ball is rotated between the poles. Not only does this permanent magnetization of the steel ball give irregularities in readings but also there is a nonuniformity of the steel ball due to its method of manufacture. This last difficulty will be met in any type of test for hardness which may be given to a

[3] S. R. Williams, TRANSACTIONS, American Society for Steel Treating, Vol. 5, 1924, p. 367.

steel ball. However, in the magnetic test which is herewith described we have a test which is not a check on any particular point of the specimen but rather of the steel ball as a whole, it is an integral effect, and for that reason seems to me a better type of test to apply. It would not be applicable to testing out local imperfections of the steel balls. A hardness testing machine like

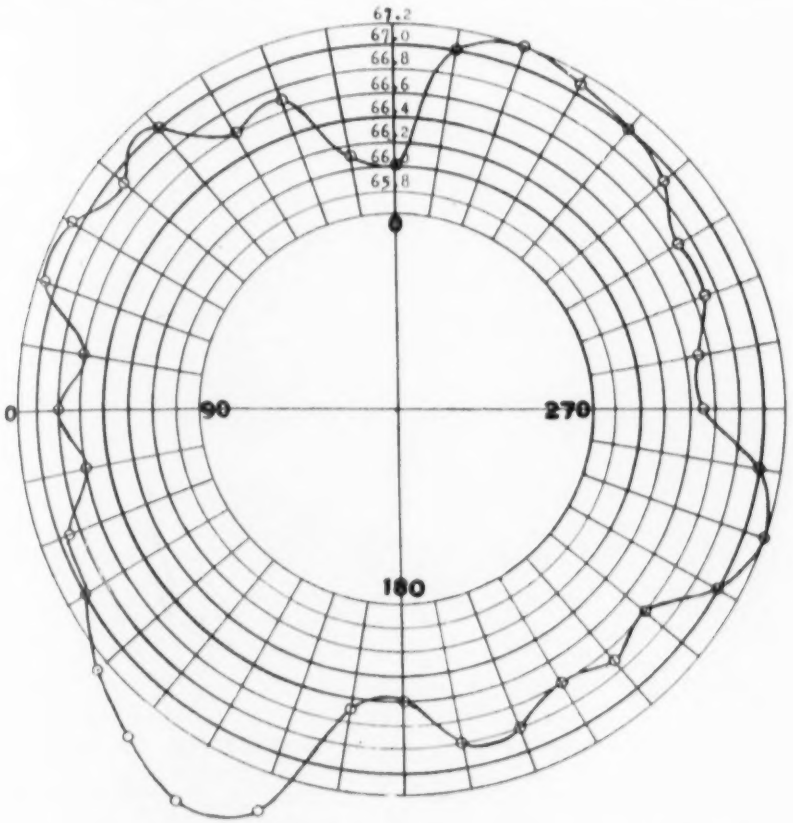


Fig. 5—Graph Showing Rockwell Hardness Numbers Taken at Points Every Ten Degrees of Arc Around a Steel Ball. The Ball is ¾ Inch in Diameter.

the Rockwell hardness tester would show up highly localized imperfections better than the magnetic method. As indicating this nonuniformity of the steel ball a series of readings were taken around one of the steel balls every ten degrees of arc and are plotted in Fig. 5. It is obvious from this diagram that there are definite polarities of hardness to steel balls and which appear to vary from ball to ball.

Fig. 6 shows the surface of break in a steel ball cracked by a blow. There appear to be definite lines to the break as though

there was a definite polarity to structure and whether it is a chance break or not these lines are continuous with magnetic lines as indicated by the attached filings. Both Figs. 6 and 7 would indicate nonhomogeneity which would show up on the magnetic hardometer. Table I gives the numerical values for the set of readings shown in Fig. 5 and for a set taken at right angles to the first.

Table I
Rockwell Hardness Values as Plotted in Fig. 5

Angle	Rockwell Numbers	Angle	Rockwell Numbers
0	66.0	180	66.4
10	66.1	190	66.8
20	66.7	200	66.8
30	66.6	210	66.6
40	67.0	220	66.7
50	66.9	230	66.6
60	67.1	240	67.0
70	67.1	250	67.2
80	66.6	260	67.0
90	66.8	270	66.5
100	66.6	280	66.5
110	66.9	290	67.0
120	67.0	300	67.0
130	67.3	310	66.9
140	67.5	320	67.0
150	67.7	330	67.1
160	67.5	340	67.2
170	66.5	350	67.0

Table I—Continued
Set of readings at right angles to those just given.

Angle	Rockwell Numbers	Angle	Rockwell Numbers
0	66.0	180	66.4
10	66.1	190	66.8
20	66.7	200	66.8
30	66.6	210	66.6
40	67.0	220	66.7
50	66.9	230	66.6
60	67.1	240	67.0
70	67.1	250	67.2
80	66.6	260	67.0
90	66.8	270	66.5
100	66.6	280	66.5
110	66.9	290	67.0
120	67.0	300	67.0
130	67.3	310	66.9
140	67.5	320	67.0
150	67.7	330	67.1
160	67.5	340	67.2
170	66.5	350	67.0

Both of these sets of readings were taken on ball No. 1 having had a tempering temperature 480 degrees Fahr. (250 degrees Cent.) and diameter = ¾ inch.

It must be remembered that the magnetic test includes the interior condition of the ball also, while the Rockwell test is one

Fig. 6—Photograph of the Surface of a Break in a Hardened Steel Ball Split by a Hammer.

of surface layer conditions only. Unless the steel ball were homogeneous it would not be possible to work out a satisfactory set of conversion factors between Rockwell and magnetic numbers.

One of the first tests carried out was a series of readings on a set of steel balls, ¾ inch in diameter, which had different tempering temperatures and these readings were then compared with the rebound tests and the Rockwell numbers. The balls with different tempering temperatures were used in the previous

paper and were, therefore, demagnetized before testing. They were not demagnetized in the former paper and so the results cannot be compared. There were three balls to each temperature, of which there were four different temperatures, viz., 480, 660, 840, and 1020 degrees Fahr. (250, 350, 450, and 550 degrees Cent.), respectively. The readings on the three different balls for each temperature were averaged together for the values given in Fig. 7, but in Table II the individual readings for the different balls are indicated and will give some idea of the way in which the magnetic numbers run on the magnetic hardometer.

Table II
Magnetic Numbers

	Tempering Temperature, 480°F.				
Ball No. 1	65.0	65.6	65.5	65.4	
Ball No. 2	65.5	65.7	65.0	65.4	Average value 65.0
Ball No. 3	64.3	64.3	63.9	64.2	
	Tempering Temperature, 660°F.				
Ball No. 1	61.5	61.0	61.0	61.2	
Ball No. 2	60.0	61.5	61.0	60.8	Average value 61.1
Ball No. 3	61.9	61.3	61.1	61.4	
	Tempering Temperature, 840°F.				
Ball No. 1	53.7	52.3	52.5		
Ball No. 2	54.0	53.3	53.2		Average value 53.2
Ball No. 3	53.3	53.2	52.9		
	Tempering Temperature, 1020°F.				
Ball No. 1	53.5	53.3	52.3		
Ball No. 2	53.3	53.0	52.8		Average value 52.9
Ball No. 3	53.2	52.5	52.3		

A series of readings was then taken on these same balls by the rebound method. The striking surface for the balls was the polished surface of a very hard anvil. On the various surfaces thus tried out the old anvil with its surface smoothed off gave the best results. The rebound numbers are tabulated in the accompanying table.

Table III
Rebound numbers = Height of rebound in centimeters
Dropping height = 70cm

	Tempering Temperature, 480°F.			
Ball No. 1	66.2	66.3	66.2	
Ball No. 2	66.0	66.5	66.5	Average number 66.3
Ball No. 3	66.2	66.5	66.3	
	Tempering Temperature, 660°F.			
Ball No. 1	65.8	65.6	66.0	
Ball No. 2	65.9	65.7	66.0	Average number 65.8
Ball No. 3	65.5	65.8	66.1	

Table III—(Continued)

	Tempering Temperature, 840°F.			
Ball No. 1	64.6	64.8	64.8	
Ball No. 2	64.5	64.2	64.2	Average number 64.6
Ball No. 3	64.6	64.8	64.9	
	Tempering Temperature, 1020°F.			
Ball No. 1	62.0	62.5	62.6	
Ball No. 2	61.9	62.2	62.1	Average number 62.4
Ball No. 3	62.5	62.6	62.6	

The same series of steel balls was next put to the test of the Rockwell hardness tester. Inasmuch as this instrument left in-

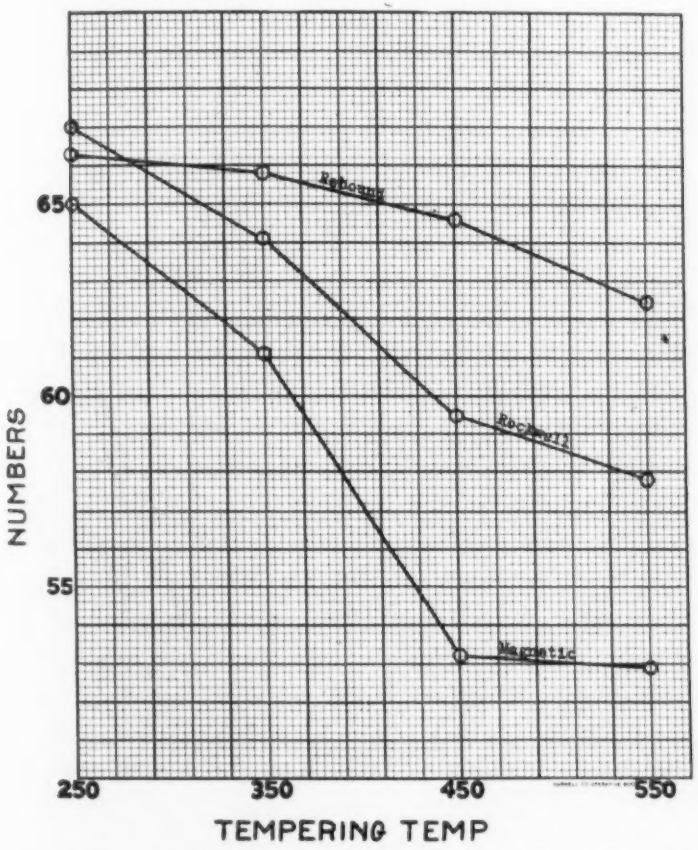


Fig. 7—Curves Showing a Comparison of a Rebound, Rockwell and Magnetic Numbers.

dentations in the balls which would influence magnetic tests, only one ball was taken from each tempering temperature. The readings for the different Rockwell numbers were taken at random around the ball and are given in Table IV. The balls selected were those showing the largest differences in magnetic numbers as recorded in Table II.

A comparison of the three sets of readings is shown in Fig. 7. All of the readings are compared with the tempering temperatures. The Rockwell numbers are a measure of indentation hardness and express this property in terms of plasticity while the rebound numbers measure the elastic properties of the metal. The question may properly be raised, what do the magnetic numbers measure?

Table IV
Rockwell Numbers

Tempering Temperature 480°F. Ball No. 1	Tempering Temperature 660°F. Ball No. 2	Tempering Temperature 840°F. Ball No. 2	Tempering Temperature 1020°F Ball No. 3
67.3	64.2	59.6	57.7
66.6	64.2	59.6	57.9
67.0	64.2	59.8	58.0
67.0	64.2	60.1	58.0
66.8	64.0	59.7	57.7
67.1	64.0	60.0	57.6
66.9	64.5	59.4	57.9
66.8	63.6	59.3	57.8
67.1	64.4	59.6	57.9
66.9	64.5	59.3	57.5
67.0	63.9	59.5	57.9
67.7	64.3	59.5	58.1
67.0 Av.	64.1 Av.	59.5 Av.	57.8 Av.

Steel balls are put on the market with a fairly definite heat treatment and for any particular brand should be quite uniform. It was of interest, therefore, to take a hundred steel balls, supposedly of the same origin and treatment, and see if the magnetic hardometer would differentiate them with respect to hardness. On a scale which could be moved into any position desired it was arranged that those balls which gave a reading between 73 and 69 would be classed as very hard, between 69 and 65 as hard, 65 to 61 as medium, 61 to 57 as soft and below 57 as very soft. The distribution of these balls according to hardness follows roughly a general distribution law. The values for the rebound numbers are also given.

Table V

	Very hard	Hard	Medium	Soft	Very soft
No. of Balls	3	17	24	50	6
Magnetic Numbers	73-69	69-65	65-61	61-57	57-
Rebound Numbers	66.0	60.9-63.5	55.9-58.4	55.9-60.9	55.9-60.9

This set of balls was needed for further work and so the Rockwell hardness tester could not be applied to all of them. Two were selected, according to the magnetic numbers, one from the very hard and one from the very soft. The record for these two are given in Table VI. Values were taken 20 degrees apart in a circle around the balls.

Table VI

Very Hard Ball		Very Soft Ball	
Angles	Rockwell Numbers	Angles	Rockwell Numbers
0	65.8	0	63.4
20	66.3	20	65.2
40	66.5	40	65.1
60	66.4	60	65.2
80	65.8	80	65.3
100	65.7	100	65.3
120	66.2	120	65.2
140	66.0	140	65.3
160	66.1	160	65.3
180	66.3	180	64.9
200	66.0	200	65.3
220	66.2	220	65.2
240	66.1	240	65.3
260	66.3	260	65.2
280	66.3	280	65.0
300	66.5	300	65.0
320	66.0	320	65.2
340	66.1	340	65.2
	Av. 66.2		Av. 65.2

Both Tables V and VI indicate that whatever property the magnetic numbers measure they are making that distinction very clear. In the previous paper it was shown that those steel balls which were classified as uniform and hard by the magnetic tester gave the best record for crushing strength. In all cases the magnetic numbers seem to go progressively with those numbers which in other methods we call the hardness numbers of the material tested. Not only do they follow some law of comparison but the different degrees of hardness are indicated by larger steps in the magnetic numbers than by other methods which makes it an easy method to use. It is not clear just what the kind of hardness is which the magnetic hardometer measures, but Fig. 7 would indicate that it comes nearer to the Rockwell hardness numbers than to the others. Can we say that the magnetic numbers measure the resistance to penetration? Much more work must be done on the relation between magnetism and hardness and the direc-

tion of this line of work I hope to show in another paper. One cannot speak of the numbers which are obtained by this method as the magnetic hardness of the material, for, unfortunately, even from the magnetic viewpoint, the term "magnetic hardness" has been very frequently used to indicate the coercive force for saturation.[4] The use of such a term in our present problem would lead to misunderstandings although it would more properly be used in that connection than with coercive force. It may develop that coercive force and hardness as measured by a hardness tester do have a definite relation. Work by Nusbaum[5] indicates this and some work which the author now has under way may throw more light on the subject.

Another item of interest in comparing Rockwell and magnetic numbers was a series of readings with the Rockwell instrument across the surface of a section taken through the center of the ball. Such a set of readings would show whether the outer layers were harder than the interior. Table VII gives these readings taken at about equal distances across the section of a $\frac{7}{16}$ inch ball.

Table VII
Rockwell Numbers Along Diameter of $\frac{7}{16}$-Inch Steel Ball

Readings	Rockwell Numbers
No. 1	66.0
No. 2	66.5
No. 3	66.8
No. 4	66.4
No. 5	66.2
No. 6	66.6
No. 7	66.3

This set of numbers does not reveal any striking difference between the hardness numbers for the outer layers and the center.

The method as thus presented seems to offer possibilities for picking out the high grade steel balls without injuring them and it is hoped that some manufacturing plant will try it out in regular routine testing. The outfit is inexpensive and may be built in any laboratory. *This investigation has its chief value for the present in this that it gives another angle from which to view the property called hardness as measured by well known methods.*

[4]Wild, *Transactions*, Faraday Society, 9-11, 1913-16.

[5]Nusbaum, *Proceedings*, American Society for Testing Materials, Vol. 19, 1919, Part 2.

(Continued on Page 823)

GRAPHITIZING BEHAVIOR OF IRON CARBIDE IN PURE IRON-CARBON ALLOYS IN THE CRITICAL RANGE

By H. P. Evans and Anson Hayes

Abstract

This paper presents a theoretical treatment of the action of carbon monoxide—carbon dioxide mixtures in catalyzing the graphitization of iron carbide in iron-carbon alloys. The experimental proof is presented that iron carbide in iron-carbon alloys of rather high purity and containing about 2.30 per cent carbon is metastable at temperatures below 1292 degrees Fahr. (700 degrees Cent.) and at temperatures above 1562 degrees Fahr. (850 degrees Cent.) in the presence of this gas mixture when it is applied at a pressure of 5 atmospheres.

Review of Literature

THE three double-line iron-carbon diagrams in most general use are those of Benedict[1], of Ruff[2] and of Wittorf[3]. The portion of these three diagrams, which is important in explaining the graphitizing properties of pure iron-carbon alloys below the solidification range, i. e., below 2073 degrees Fahr. (1134 degrees Cent.), differ only in the relative slopes of the carbon solubility and of the cementite solubility lines. (The latter is the A_{cm} line).

Benedict's diagram (Fig. 1) shows the slope of the A_{cm} line to be greater than that of the carbon solubility line. Wittorf indicates that the two lines are parallel, while Ruff has drawn them with the carbon solubility line having the greater slope. Benedict's diagram would require also, as would Wittorf's as well, that pure iron-carbon alloys should graphitize completely on cooling slowly, since the carbon solubility line lies to the left of the cementite solubility line. Ruff's diagram, however, offers the possibility of an intersection, and for temperatures below it, iron carbide should be stable.

[1] *Metallurgie*, 5, page 45, 1908.

[2] *Metallurgie*, 8, page 457, 1911.

[3] *Journal*, Russian Physical Chemical Society, 43, 1613, 1911. (Original article not read.)

A paper presented before the eighth annual convention of the Society, Chicago, September 20 to 24, 1926. Of the authors, Dr. Anson Hayes is professor of physical chemistry and Dr. H. P. Evans instructor in chemistry, Iowa State College, Ames, Iowa.

It is well known that iron carbide in white cast iron for malleable purposes is metastable throughout the entire critical range. In the case of pure iron-carbon alloys, however, this fact has not been established. At temperatures slightly below 2073 degrees

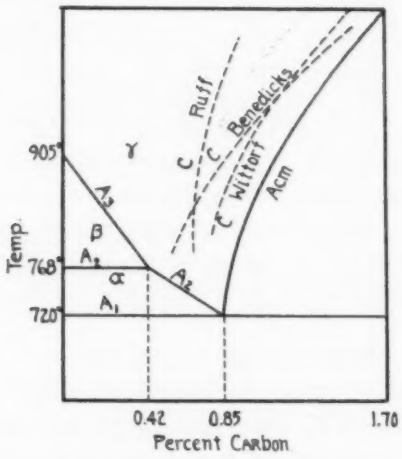


Fig. 1—Iron-Carbon Equilibrium Diagram.

Fahr. (1134 degrees Cent.) iron carbide is known to be metastable. Honda states, however, that pure iron-carbon alloys will not graphitize below 1652 degrees Fahr. (900 degrees Cent.). Schwartz has held a sample at 1652 degrees Fahr. (900 degrees Cent.) for over 2000 hours with no indications of graphitization. A search of the literature yields no report of graphitization below 1652 degrees Fahr. (900 degrees Cent.)

Maxwell and Hayes, however, in work done in this laboratory[1], established the fact that iron carbide is metastable at temperatures of 1202 and 1292 degrees Fahr. (650 and 700 degrees Cent.). These facts would indicate either that iron carbide is stable in the temperature interval 1292 to 1652 degrees Fahr. (700 to 900 degrees Cent.), or that if metastable, it exists in a very persistent metastable state.

Theoretical Discussion

The results obtained were for iron-carbon alloys containing small amounts of impurities. The analysis of the alloys accompanies the results. While no claim is made that the alloys are pure, the very low value for the impurities present should produce behaviors

[1] *Journal*, American Chemical Society, Vol. 48, page 584, 1926.

approaching very closely that of pure iron-carbon alloys. With these facts in mind, consideration of these alloys will be resumed, calling them pure iron-carbon alloys for the sake of convenience.

It was evident at the outset of the investigation that iron carbide, if metastable in this temperature range existed in a very persistent metastable condition. An attempt was made to establish the optimum conditions for the breaking down of iron carbide possibly by exposing it to the action of some catalytic agent which would hasten the rate of decomposition. Hayes and Scott[5], in work done in this laboratory, showed that in the case of white cast iron at a temperature of 1701 degrees Fahr. (927 degrees Cent.), carbon monoxide–carbon dioxide mixtures at a pressure of 5 atmospheres increased the rate of absorption of free iron carbide nearly 100 per cent.

The following reactions may be considered to accurately represent a possible mechanism for the graphitization of pure iron-carbon alloys below the critical range.

$$(1)\quad 2\,CO = C + CO_2$$
$$(2)\quad 3Fe\ (alpha) + 2CO = Fe_3C + CO_2$$
$$(3)\quad Fe_3C = 3\,Fe\ (alpha) + C$$

The fact was pointed out by Schenck that the net result of reactions (1) and (2), i. e. reaction (1) proceeding toward the right and reaction (2) toward the left, is the equivalent of reaction (3) proceeding toward the right.

Now if we let $\triangle F_1$, $\triangle F_2$, and $\triangle F_3$ represent respectively the change in the free energy content (the change in the value of the Gibbs Zeta function, for a more complete discussion of which see "Thermodynamics" by Lewis and Randall, McGraw-Hill) accompanying each of these reactions when proceeding from left to right then the second law of thermodynamics demands that $\triangle F_1 - \triangle F_2 = \triangle F_3$. If this were not the case, a cyclical process involving reaction (3) proceeding in one direction and of reactions (1) and (2) so as to carry out the equivalent of reaction (3) in the opposite direction, could be carried out in such a manner as to accumulate an indefinite quantity of available energy. Thus, to be more specific, assume that reaction (3) represents a spontaneous process when proceeding from left to right and that during the breakdown of one formula weight of the carbide, a quantity of available

[5]American Foundrymen's Association, vol. 33, p. 574.

energy $\triangle F_3$ is stored up and that this quantity of energy is greater than $\triangle F_1 - \triangle F_2$, which latter quantity represents the available energy stored up when the same change is carried out by means of reactions (1) and (2).

Under such a set of conditions, it would be possible to use a portion of the energy $\triangle F_3$ to produce a formula weight of the carbide Fe_3C from carbon and alpha iron by means of the two reactions (1) and (2) carried out in the reverse direction. When this is accomplished, all of the various substances are in the original condition and as a net result we have gained a quantity of available energy represented by the quantity $\triangle F_3 - (\triangle F_1 - \triangle F_2)$. The conclusion thus reached is that $\triangle F_3 = (\triangle F_1 - \triangle F_2)$.

The mass law equilibrium constants for reactions (1) and (2) may be represented by the following expressions:

$$K_1 = \frac{P_{1-CO_2}}{P^2_{1-CO}} \quad \text{and} \quad K_2 = \frac{P_{2-CO_2}}{P^2_{2-CO}}$$

where P represents the partial pressures of the carbon monoxide and carbon dioxide at equilibrium with the solid phases for reactions (1) and (2) respectively.

Now $\triangle F_1 = -RT\ln K_1$ and $\triangle F_2 = -RT\ln K_2$ where the reactants are brought into the reaction at unit pressure and where the products of the reaction are removed at unit pressure (See Lewis and Randall, "Thermodynamics", McGraw-Hill), and where R is the molal gas constant and T is the absolute temperature.

From these relations it follows that

$$\triangle F_3 = \triangle F_1 - \triangle F_2 = RT(\ln K_2 - \ln K_1).$$

If reaction (3) is a spontaneous process $\triangle F_3$ must be negative. i. e., the free energy content of Fe_3C must be greater than that of 3 Fe(alpha) + C. This in turn requires that

K_2 is less than K_1 or that $\frac{P_{2-CO_2}}{P^2_{2-CO}}$ is less than $\frac{P_{1-CO_2}}{P^2_{1-CO}}$ or

if the total pressure is constant at one atmosphere, this requires that $P_2{-}CO_2 < P_1{-}CO_2$ and that $P_2{-}CO > P_1{-}CO$.

Above the critical range Fig. 2, which is taken from Schenck and which represents the results of equilibrium studies of Matsubara, requires that the carbide in pure iron-carbon alloys be stable.

Attention should be called to the fact that at these temperatures, the partial pressure of carbon monoxide is higher in equilibrium with carbon than it is in contact with saturated austenite and free iron carbide. Also, the partial pressure of carbon dioxide is greater

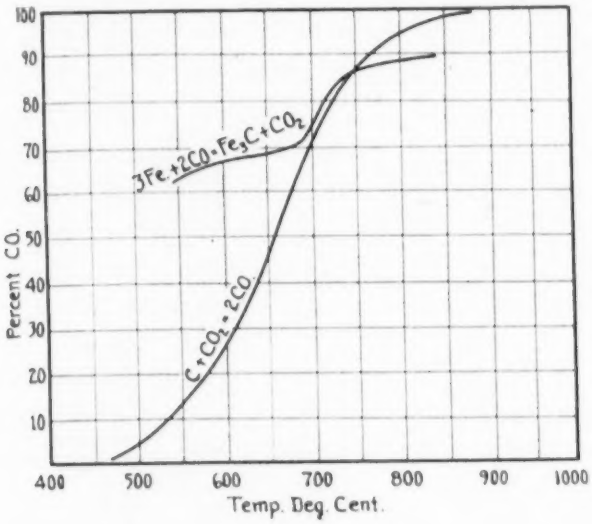


Fig. 2—Gas Equilibrium Diagram.

in contact with iron carbide and austenite than with carbon.

That this condition of pressure gradient exists when iron carbide is the stable phase in pure iron-carbon alloys may be shown from a consideration of the changes of state necessary for the formation of one formula weight of Fe_3C from the gamma iron of austenite and carbon as follows:

(1) $2\,CO = C + CO_2$

Free energy change $= \triangle F_1 = -RT\ln K_1$.

(2) 3Fe (gamma in Fe_3C austenite) $+ 2\,CO = Fe_3C + CO_2$
Free energy change $= \triangle F_2 = -RT\ln K_2$.

(3) 3Fe (gamma in C austenite) $=$ 3 Fe (gamma in Fe_3C austenite).
Free energy change $= \triangle F_3$

Addition of (2) and (3) minus (1) gives

(4) 3Fe (gamma in C austenite) $+ C = Fe_3C$.

Free energy of formation of iron carbide $= \triangle F_4$.

Or, $\triangle F_4 = \triangle F_2 - \triangle F_1 + \triangle F_3 = RT\,(\ln K_1 - \ln K_2) + \triangle F_3$.

Now if Fe_3C is the stable phase, the free energy content of 3Fe (gamma in C austenite) plus the free energy content of 1 C (in C austenite) is greater than the free energy content of one Fe_3C in

Note—The expression "(C austenite)" and "(Fe_3C austenite)" are used to represent austenite saturated with carbon from carbon and from Fe_3C respectively.

the free state, i. e. $\triangle F_4$ is negative. Furthermore, $\triangle F_3$ must be positive. This latter conclusion may be reached as follows.

The solubility of a constituent from a metastable system is always higher than from a stable phase. If Fe_3C is the stable phase, austenite saturated with carbon from carbon contains more carbon than does austenite saturated with carbon from Fe_3C. This being the case, the concentration of gamma iron (in C austenite) would

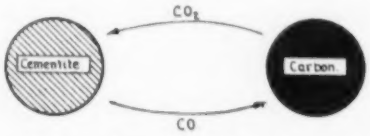


Fig. 3—Gas Mixture Acting as a Carrier of Carbon in Graphitizing Process.

be less than it is in (Fe_3C austenite). Since energy is required to transfer a constituent from a lower to a higher concentration, $\triangle F_3$ is positive if Fe_3C is the stable phase. Going back to a consideration of the relation

$$\triangle F_4 = \triangle F_2 - \triangle F_1 + \triangle F_3 = RT\,(\ln K_1 - \ln K_2) + \triangle F_3$$

the following conclusion may be reached.

$\triangle F_4$ is negative since Fe_3C is assumed to be the stable phase and $\triangle F_3$ is positive, therefore $\triangle F_1$ must exceed the sum of

$$\triangle F_2 + \triangle F_3 \text{ or}$$
$$\triangle F_1 > \triangle F_2 \text{ and}$$
$$K_1 < K_2 \text{ or}$$
$$\frac{P_{1-CO_2}}{P^2_{1-CO}} < \frac{P_{2-CO_2}}{P^2_{2-CO}}$$

from which it will be seen that a pressure gradient of CO should exist from carbon to Fe_3C and of CO_2 in the opposite direction.

Considerations exactly similar to these will show that if Fe_3C

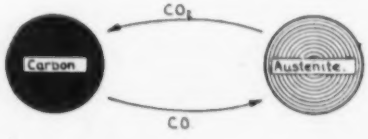


Fig. 4—Gas Mixture Acting as a Carrier of Carbon in Carburizing Process.

is assumed to be the metastable phase, K_2 is less than K_1 or the direction of the pressure gradients are reversed.

From the above treatment it is to be expected that if iron containing carbon near the saturated value is permeable to a gas mixture of carbon monoxide and carbon dioxide, these gases should assist in the graphitizing process if Fe_3C is metastable. Fig. 3 shows the gas mixture acting as a carrier of carbon from cementite to carbon spot.

A consideration of the mechanism by which this gas mixture acts will show definitely that a gas mixture richer in carbon monoxide than that corresponding to the equilibrium value for carbon will introduce carbon into austenite saturated with carbon from carbon. See Fig. 4. It also indicates that a still higher value of carbon monoxide would convert iron into free iron carbide and that

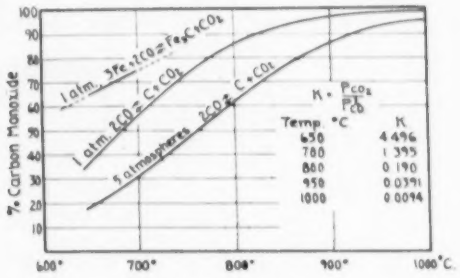


Fig. 5—Compositions of Gases in Equilibrium with Carbon and in Equilibrium with Alpha Iron and Iron Carbide.

a carbon monoxide content less than that corresponding to equilibrium with carbon would oxidize carbon from a sample saturated with carbon either from cementite or carbon.

If, then, a pure iron-carbon alloy be subjected to a gaseous mixture, keeping on the carbon dioxide side of equilibrium values, it is to be expected that at some depth in the piece the gas mixture will reach an equilibrium value with carbon. In the exterior of the piece the excess carbon dioxide will act as an oxidizing agent, oxidizing iron carbide to iron and carbon monoxide. As the gas mixture diffuses into the piece, the carbon dioxide in contact with iron carbide will continue to pick up carbon, by forming carbon monoxide, until at a certain depth the gas mixture reaches an equilibrium value with carbon. Immediately beyond this point the gas mixture should catalyze the breakdown of cementite and the formation of carbon according to Fig. 3.

The following experimental work was carried out in an attempt to get more definite information in regard to the stability

of iron carbide in the temperature interval from 1472 degrees Fahr. (800 degrees Cent.) to those where the work of other investigators has definitely shown the metastable character of this compound. In Fig. 5 is shown the carbon monoxide content of the carbon monoxide and carbon dioxide mixtures in equilibrium with carbon over the temperature range 1202 to 1832 degrees Fahr. (650 to 1000

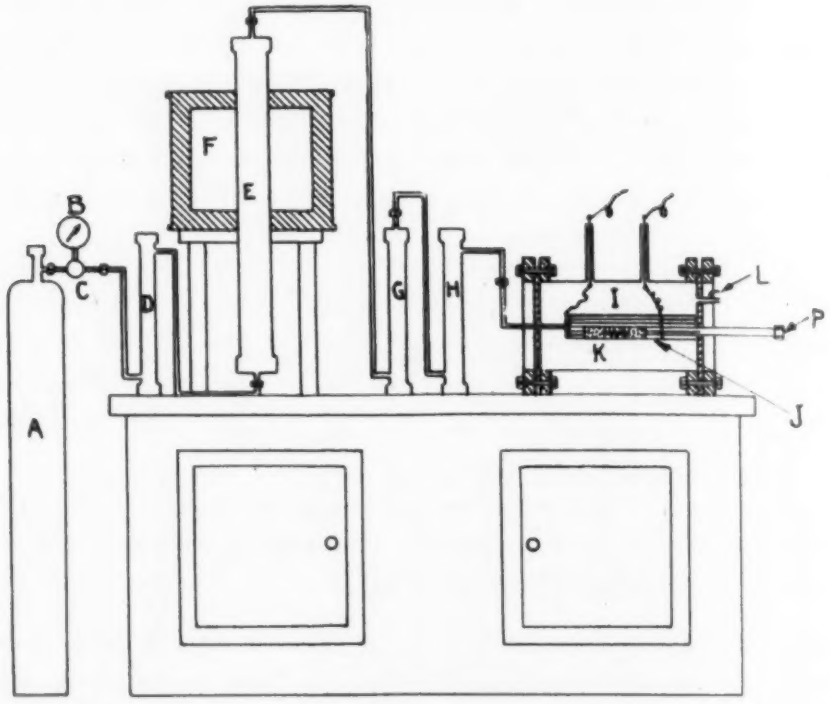


Fig. 6—Apparatus for Annealing Under Various Pressures of Carbon Monoxide, Carbon Dioxide Mixtures. A—Carbon Dioxide Tank, B and C—Pressure Regulator, D and H—Calcium Chloride Drying Towers, F—Muffle Furnace, Set on End, E—Iron Pipe, Packed with Carbon, G—Potassium Hydroxide Tower, I—Silocel, K—Sample, Packed in Graphite, J—Heating Element, L—Gas Cock, P—Observation Tube.

degrees Cent.) and at pressures of 1 and 5 atmospheres as given in data from work of Rhead and Wheeler[6].

The values for temperatures below 1472 degrees Fahr. (800 degrees Cent.) were obtained by extrapolation from the higher temperatures where experimental determinations were obtained by Rhead and Wheeler. In this extrapolation the change in the heat content $\triangle$ H of the reaction

$$2\,CO = C + CO_2$$

was taken as —41950 calories and the equation

[6] *Journal*, American Chemical Society, Vol. 99, page 1140, 1911.

$$2303 \log \frac{K_2}{K_1} = \frac{\triangle H}{R}\left[\frac{1}{T_1} - \frac{1}{T_2}\right]$$

was used.

Apparatus and Method of Procedure

The apparatus shown in Fig. 6 was developed in which gas mixtures of varying ratios of carbon monoxide and carbon dioxide contents could be obtained and in which those mixtures could be applied to the sample at pressures up to 100 pounds per square inch and at temperatures up to 1832 degrees Fahr. (1000 degrees Cent.). A description of this apparatus and the manner of operating it follows.

A is a carbon dioxide tank, B is a pressure gage and C a control valve. D and H are drying chambers made from two inch black pipe eighteen inches long, threaded at both ends and fitted with caps. These chambers are filled with calcium chloride and phosphorus pentoxide. The carbon chamber E is made from three inch black pipe, three feet long, threaded at both ends and fitted with caps. This chamber is filled with carbon. The lower end is filled with loose gas carbon and the part exposed inside the muffle furnace F is packed with powdered wood charcoal or graphite. A thin layer of gas carbon is packed on top of this. This tube is filled only to the point at which it emerges from the furnace at the top. G is a chamber, similar in size and construction to D and H and in some preliminary runs was filled with a 35 per cent solution of potassium hydroxide to absorb any great excess of carbon dioxide. I is the pressure furnace. It is made from a piece of ten inch steam main, eighteen inches long, threaded at both ends and fitted with companion flanges. The ends are closed by bolting copper plates between the flanges at each end. A rubber gasket between the plate and the inside flange serves to make the chamber hold pressure up to 100 pounds. In the center of this chamber is placed a twelve inch heating element J. This is held in place by a silocel packing and heat insulator. The gas mixture is led in through the rear plate directly to the end of this element. The end of the element is closed with "transite" board washers. The element is heated by 110-volt alternating current and the temperature controlled by placing rheostats in the circuit. K is a nickel receptacle with an iron lining on its upper surface. The sample is placed in this and packed with finely divided carbon, graphite or wood char-

coal, leaving only the end next to the observation tube P exposed. The temperature is recorded by means of an optical pyrometer focused through P on the sample and is also checked by an iron constantan couple which was inserted through the rear plate of the furnace. This couple was attached to an automatic recorder. The stopcock L is left partially open during the anneal and the gas mixture burned at this point. At frequent intervals samples of the gas were withdrawn here, collected over mercury, and analyzed for carbon monoxide and carbon dioxide by use of an Orsat gas analysis apparatus.

The sample to be annealed was placed in K and packed with carbon as described. The receptacle was placed approximately in the center of the heating element and the front plate and flange bolted on. Transite board washers with a one-fourth inch hole were used to fill the end of the heating element. They prevent considerable radiation and are specially necessary in annealing long samples.

As soon as the proper pressure was reached the current was turned on in the pressure furnace without any resistance in the circuit when starting. It required about fifteen minutes for the sample to come to dull red heat at this rate, and about twenty minutes longer to reach 1701 degrees Fahr. (927 degrees Cent.). By adjustment of the rheostats this temperature can be accurately maintained for any given length of time. On cooling it requires twenty-five to thirty minutes to cool from 1701 degrees Fahr. (927 degrees Cent.) down to dull red heat again.

Preparation of Alloys

Pure iron-carbon alloys were made by melting Armco iron and pure Acheson graphite. The iron was cut into discs from rods, and the graphite was finely divided. Weighed quantities of the iron discs and graphite were placed alternately in a magnesia-lined graphite crucible and melted in an induction furnace. The magnesia linings were made from highest purity magnesium oxide. The alloy was held in the molten state for one-half hour to insure uniform composition. The alloys prepared weighed approximately 100 grams and on microscopic examination proved to be free from graphitic carbon. A representative analysis of the alloys gave the following composition:

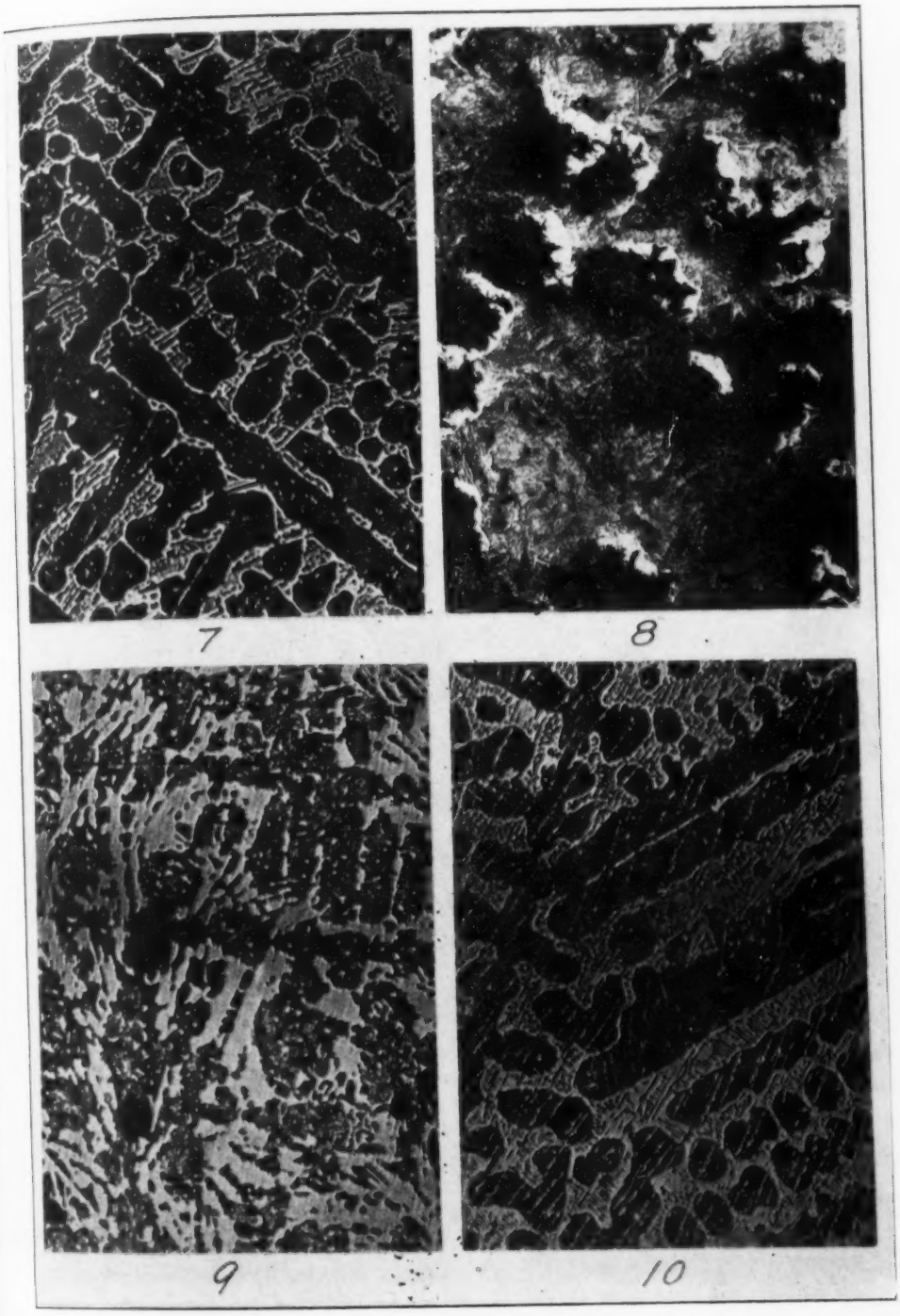


Fig. 7—Sample 1, Original. Etched in Nitric Acid. 90x. Fig. 8—Sample 1, Depth $\frac{3}{16}$-Inch. Etched in Nitric Acid. 90x. Fig. 9—Sample 1. Center of Piece. Etched in Nitric Acid. 90x. Fig. 10—Sample 2, Original. Etched in Nitric Acid. 90x.

Sulphur	Silicon	Phosphorus	Manganese	Carbon
0.025%	0.022%	0.0045%	0.045%	2.34%

Experimental Data and Discussion

Sample I: The original sample on which treatments below were made is shown in Fig. 7. It consists of free cementite and pearlite. In Figs. 8 and 9 are shown the results of a 5-hour anneal at 1701 degrees Fahr. (927 degrees Cent.) under a pressure of 5 atmospheres in a carbon monoxide–carbon dioxide mixture whose composition varied during the run from 80 to 85 per cent carbon monoxide. The samples were etched in 5 per cent nitric acid in alcohol. There was complete absorption of free cementite almost to the center of the piece. In Fig. 8 is shown the structure at a depth of approximately $\frac{3}{16}$ inches. Absorption of free cementite is complete and considerable bull's eye structure is evident. In Fig. 9 is shown the structure at the center of the piece. A comparison of this figure with the original shown in Fig. 7 shows a marked absorption of cementite and deposition of temper carbon.

Sample II: The original sample on which treatments below were made is shown in Fig. 10. This sample is similar in structure to that of Sample I. In Figs. 11 to 16 inclusive are shown the results of a 27-hour anneal at 1517 degrees Fahr. (825 degrees Cent.) $\pm$ 25 degrees under a pressure of 5 atmospheres in contact with a carbon monoxide–carbon dioxide gas mixture containing 55 per cent carbon monoxide. From Fig. 5 it will be observed that at 1472 degrees Fahr. (800 degrees Cent.) the equilibrium value for CO in a CO $-$ CO_2 gas mixture in contact with carbon under a pressure of 5 atmospheres is slightly above 60 per cent. This experiment, therefore, was carried out on the CO_2 side of equilibrium values, i. e., in a slightly oxidizing atmosphere. This sample which was in the form of a short rod, was polished on the side rather than on the end. By repeated removal of thin sections the course of graphitization could be traced into the interior of the sample.

In Fig. 11 is shown the structure toward the exterior of the piece. As is to be expected, the matrix shows hypoeutectoid structure. There is very little evidence of bull's eye structure. Carbon had been thrown out and has not been completely removed by oxidation. As the examination proceeds toward the interior of the sample, the structure approaches that of a eutectoid matrix. At

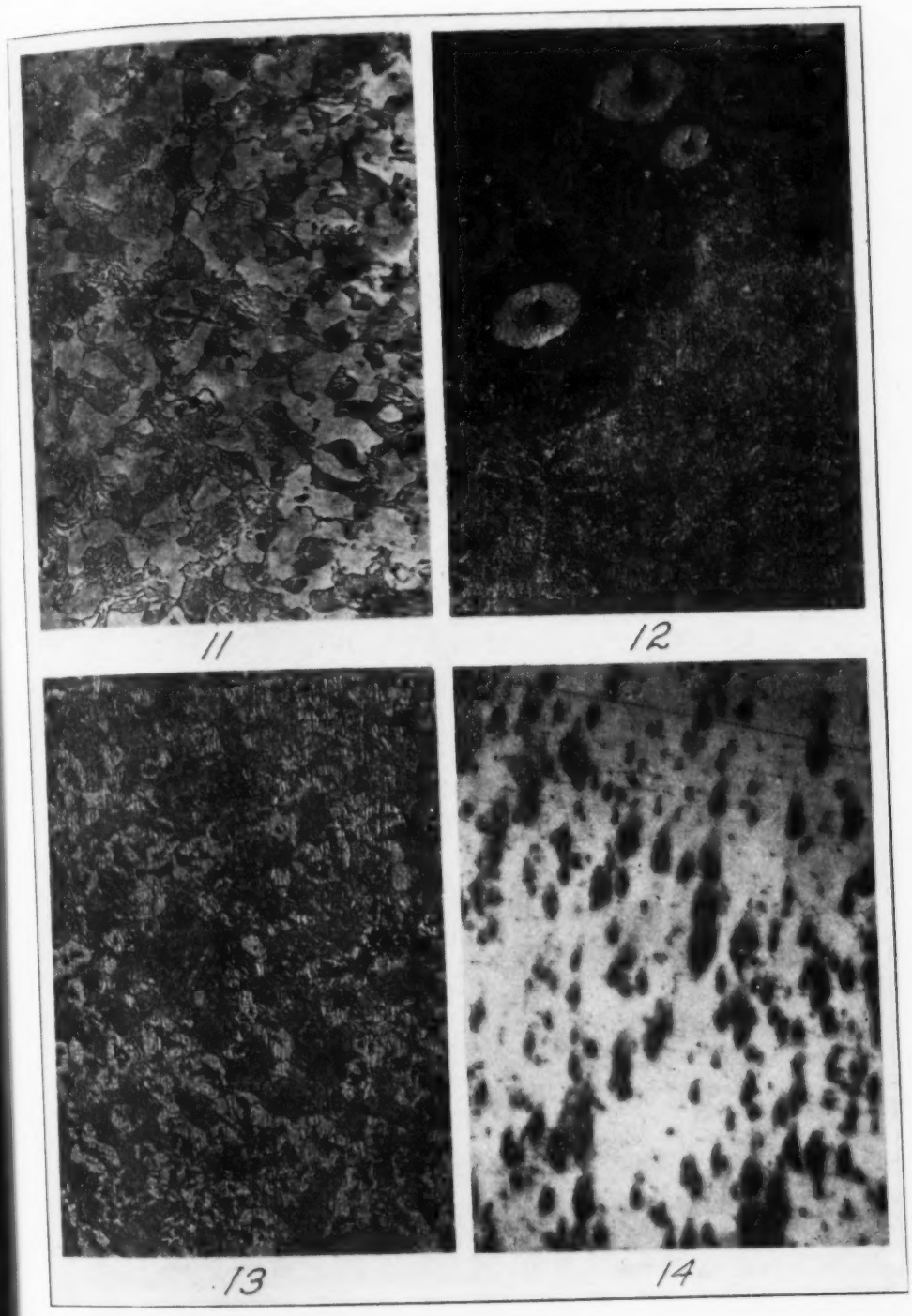


Fig. 11—Sample 2, (Close to Surface). Etched in Nitric Acid. 90x. Fig. 12—Sample 2. Depth $\frac{3}{16}$ Inch. Etched in Nitric Acid. 90x. Fig. 13—Sample 2 (Deeper than Fig. 11) Etched in Nitric Acid. 90x. Fig. 14—Same Depth as Fig. 13. Unetched. 90x.

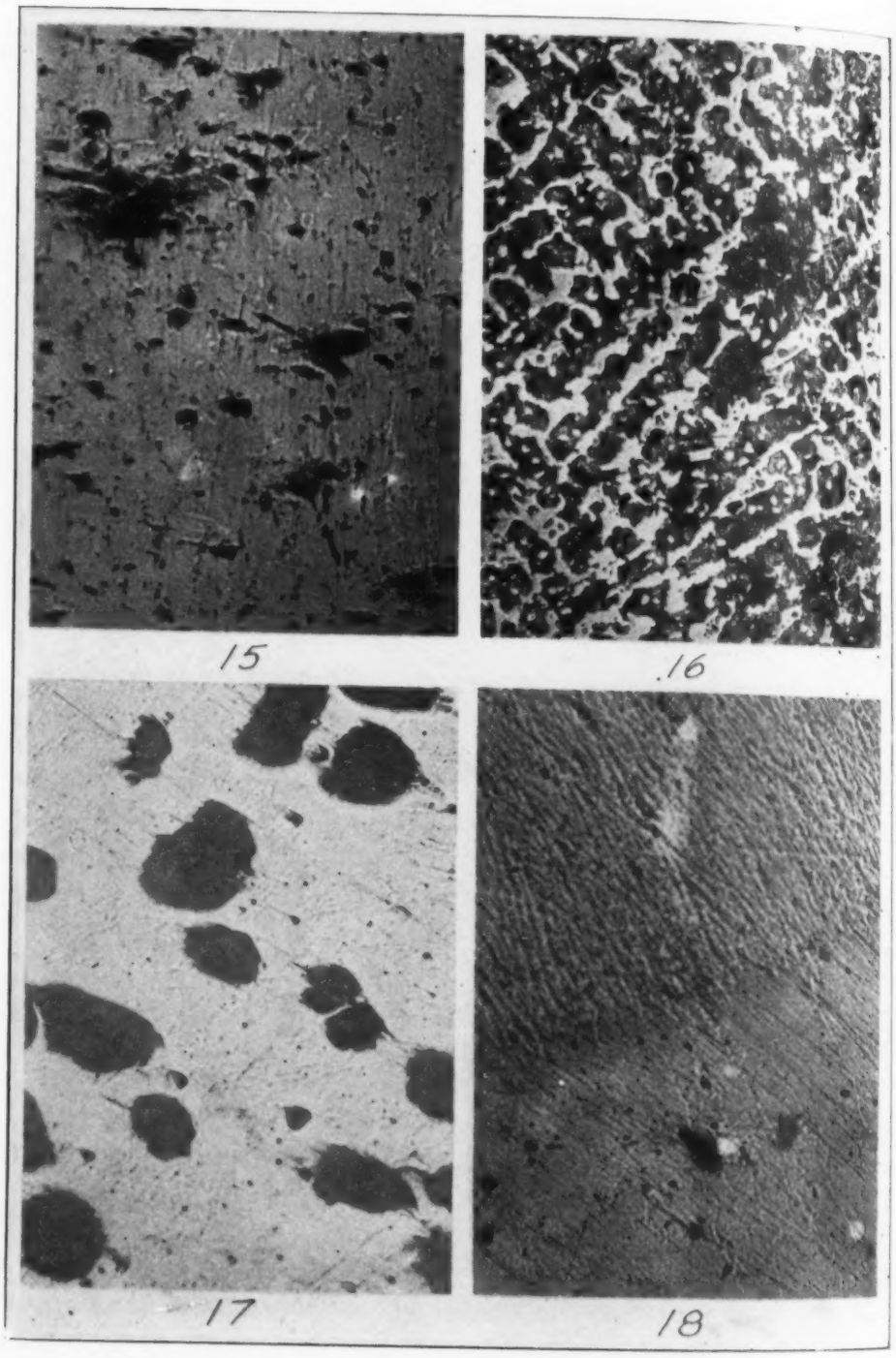


Fig. 15—Still Greater Depth. Unetched. 90x. Fig. 16—Near Center. Etched in Nitric Acid. 90x. **Fig. 17—Graphitized at 900-1000° Cent. in CO-CO_2 Mixture Under 5 Atmospheres Pressure. Fig. 18—Attempt at Graphitization at 900-1000° Cent. Packed in Carbon in Air at 1 Atmosphere Pressure.**

a depth of approximately $\frac{3}{16}$ inch the bull's eye structure, as shown in Fig. 12 has developed. The structure of the matrix at this depth is very slightly hypoeutectoid. The composition is therefore very close to that one would expect for pure iron-carbon alloys of iron-carbon eutectoid composition.

The work of Hayes and Flanders[7] on white cast iron compositions indicates that bull's eye structure is produced during the operation of the iron-carbon eutectoid. The fact that similar structures were obtained in the present work may be considered evidence of the operation of an iron-carbon eutectoid in pure iron-carbon alloys, if the temperature at which this point should occur is very near 1472 degrees Fahr. (800 degrees Cent.). Since the method of treatment is such as to make certain that at some point in the alloy compositions of austenite corresponding to that of the iron-carbon eutectoid must be produced, it is also possible that some of this bull's eye structure may have been formed while cooling the sample. In this case the conclusion would be reached that the temperature of the iron-carbon eutectoid in pure iron-carbon alloys is somewhere between 1328 and 1472 degrees Fahr. (720 and 800 degrees Cent.).

In Fig. 14 is shown the structure of the unetched sample at the same depth as Fig. 12. Abundant carbon spots are in evidence. Iron carbide is not encountered until a considerably greater depth is reached. In Fig. 15 is shown the structure in this region. Considerable unabsorbed free cementite is evident, as is also temper carbon. In Fig. 16, which is etched, is shown the structure at a still greater depth. Carbon spots are still very numerous in the pearlite areas. The structure shown in Figs. 15 and 16 continues to the center of the piece with slightly increasing free cementite areas.

Conclusion

The experimental data presented in this paper shows that iron carbide in the alloys used in this investigation is metastable at temperatures of 1562 degrees Fahr. (850 degrees Cent.) and above in the presence of a slightly oxidizing mixture of CO and CO_2, under a pressure of 5 atmospheres.

The question may be raised as to whether or not a shift from a stable to a metastable condition may have resulted from the in-

[7] Transactions of American Society for Steel Treating February, 1924, page 183.

creased pressure of 5 atmospheres. This, however, could only be the case if graphitization is accompanied with a decrease in total volume. To answer this question a piece of an alloy of the same composition was annealed under pressure of this gas mixture at a temperature near 1875 degrees Fahr. (1080 degrees Cent.). Fig. 17 shows the resulting structure. Densities were determined before and after this treatment with the following results:

Density of alloy before graphitization	Density of alloy after graphitization
7.80	7.66

Since an expansion accompanied the partial graphitization of this alloy, it is evident that the increased pressure would tend to hinder rather than to assist graphitization.

Fig. 18 shows the result of a treatment of another alloy of essentially the same composition. This treatment was carried out by packing the sample in pure graphite and in air at one atmosphere, using the same length of time and the same temperature interval.

The acceleration of graphitization brought about by the gas mixture used at 5 atmospheres is evident.

There is, however, an additional possibility. It may be that the metal absorbs sufficient quantities of CO at the high pressure to appreciably lower the chemical activity of the iron and the carbon which results from the decomposition of the iron carbide. This kind of behavior might make the carbide metastable at 5 atmospheres, and allow it to be stable in the presence of the gas mixture at one atmosphere.

In order for such a condition to exist it can be shown that rather large volumes of the gases must be dissolved in the metal. Measurements of the volume of gas absorbed at one atmosphere have been made and they are quite small. Data is not available as to the quantities of CO which are absorbed by iron at the higher pressure.

It is, therefore, only possible to say that at 1652 degrees Fahr. (850 degrees Cent.) and higher, up to about 1875 degrees Fahr. (1080 degrees Cent.) these rather pure iron carbon alloys of about 2.34 per cent carbon graphitize when in contact with suitable gas mixtures of CO and CO_2 at a pressure of 5 atmospheres.

It is worthy of note that the alloy shown in Fig. 18 which was annealed in a carbon pack in air at a temperature of 1875 degrees Fahr. (1080 degrees Cent.) shows temper carbon toward

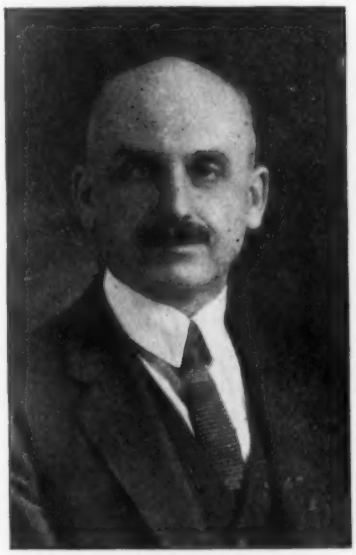
the inner portion of the decarburized rim. This is exactly what is to be expected if iron carbide is the metastable phase at these temperatures. This is exactly the same type of behavior as was obtained at temperatures below 1652 degrees Fahr. (900 degrees Cent.) when the pressures of 5 atmospheres were used.

Summary

Iron-carbon alloys of rather high purity have been graphitized in the presence of a gas mixture of CO and CO_2 at a pressure of 5 atmospheres within the temperature interval 1292 to 1875 degrees Fahr. (700 to 1080 degrees Cent.).

The Authors

Dr. Harry P. Evans received his Ph. D. degree in June, 1926, and is at present an instructor in chemistry at Iowa State College.



ANSON HAYES

H. P. EVANS

Dr. Anson Hayes is professor of physical chemistry at Iowa State College and has been directing research work in studies of the factors that influence the stability of iron carbide for the past eight years.

DISCUSSION

H. A. Schwartz: I am not familiar with all that is in Dr. Hayes' paper. From his presentation I believe that his experiment was conducted at high

pressures, and therefore, of course, this conclusively proves the metastability of the carbide. May I inquire of Dr. Hayes whether or not he obtained any information as to whether that reaction would go at all at lower pressures? Was it possible to make that graphitization at all at atmospheric pressures, or did he find the result, frequently reported, that the reaction would not start.

DR. HAYES: We made it go very readily above 900 degrees Cent.

H. A. SCHWARTZ: At atmospheric pressure of CO-CO_2?

DR. HAYES: We can get surface graphitization at lower temperatures, but I think the reason it fails to catalyze at lower temperatures is that the penetration of the metal by the gases is too slow.

H. A. SCHWARTZ: Then I would add one little thing, which is a correction to an inadvertent misquotation of a statement by myself a year or so ago. We referred to the fact that we had heated white cast iron, or, as we call it, washed metal, in a sealed quartz tube with graphite and found that no graphitization took place in 2000 hours at 1650 degrees Fahr. (900 degrees Cent.). It is not a perfectly pure alloy, but it contains some oxygen and some chromium, otherwise it is about parallel with these. The criticism was made that because the reaction did not start was no reason it did not want to start. It was not clearly understood at that time that that material case hardened until it contained 5 per cent carbon as combined carbon, which I think would dispose of the idea that no reactions wanted to get under way. I want to say that I agree thoroughly with Dr. Hayes that the cementite is metastable, but there is this one observation that bothers us as to how that phenomenon was caused. Or indeed how any carburizing with carbon occurs under this theory.

DR. S. L. HOYT: I cannot add anything experimentally, but possibly an idea or two from somebody who is outside of the field may be of some assistance, or difficulties which are suggested may be of some assistance. For instance, it was suggested that CO_2 actually goes from one place in the metal to another, and at temperatures above 900 degrees Cent. CO_2 in the free state is nonstable. It decomposes and goes practically completely over to CO at that temperature, so the question that arises in my mind is how a mechanism such as this, which is hypothecated, can actually take place.

Secondly, I gather from the presentation of the paper that this CO-CO_2 mixture accelerates or catalyzes the graphitization of cementite, and the idea occurred to me that this might possibly be accounted for on the basis of the carbon monoxide or carbon dioxide producing sufficient nuclei of graphite to start the reaction off, and the reaction once having been started by nucleation, it might then possibly proceed in the ordinary way.

DR. O. E. HARDER: I would like to inquire if Dr. Hayes, in considering this catalyzing reaction with graphite, has considered the possibility of the formation of an iron carbonyl compound? In certain work we have done on welding in atmospheres of carbon monoxide and carbon dioxide, we have had some indication of the formation of iron carbonyl. Such a compound has not been recognized in the literature, and we do not know of its existence, but I would like to know what Dr. Hayes thinks about that as a possibility in this reaction.

DR. HAYES: I have not considered the formation of iron carbonyl, neither do I know whether that equilibrium waits to be established between the gases until it gets completely over to the carbon spot from the iron carbide spot, or whether it is continually being shifted as it goes along through the austenite, which, of course, is varying in carbon as it goes through. So, as far as the harmonizing of this thing of the second law of thermodynamics is concerned, it does not matter whether the mechanism I have suggested is the one or not. It is one of the possible ones, and as long as there is a possible violation of the second law of thermodynamics I feel fairly safe in concluding that iron carbide is metastable when this gas mixture catalyzes it. I mean, it is metastable in the presence of these gases at 5 atmospheres, as I stated in the paper.

In conection with the last discussion that was made, in this work we wanted to establish the fact that the effect of 5 atmospheres pressure alone could not have shifted iron carbide from a stable to a metastable condition, and we ran experiments to show an expansion takes place when graphitization takes place, which shows it would hinder instead of promote graphitization. We did not conduct any graphitization experiments as such, below the critical range. Perhaps you noted a paper published in the Journal of the American Chemical Society in which we obtained the free energy of formation of iron carbide at 650 and 700 degrees Cent. and noted that positive free energy formation, was decreasing rapidly as we go up. Whether that might go to zero and cause it to be stable around 750 and 800 degrees Cent. is more than I am able to say. That is the reason I refrained from drawing conclusions. It is a well known fact that iron carbide will graphitize under some conditions in pure alloys and will not graphitize under others, which suggests it must be very close to the dividing line between the metastable and stable condition, and I am wondering if it is not possible that there is a reaction like this carbon dioxide reacting with iron to form iron oxide, the iron oxide dissolving into the austenite might lower the activity of iron in the austenite to such a degree as to render the iron carbide metastable in the presence of these gases when it would behave as a stable body in the absence of it. I refrain from drawing conclusions on that, also.

We did not prove as much as we had hoped to prove when we started, but we have to report that we graphitized pure iron carbon alloys under conditions that had not been reported before.

DR. S. L. HOYT: The second law of thermodynamics tells us absolutely nothing about the mechanism of a reaction. It relates only to the energy relationships that call for that reaction. Dr. Hayes suggests a mechanism of which there is only one possible mechanism. Now, as I interpret that, CO molecules must migrate through it and CO_2 molecules must migrate through the lattice. I can conceive of carbon migrating fairly readily through the lattice, but when it comes to molecules such as CO or CO_2 migrating I must admit it is beyond my conception. My question is how does Professor Hayes conceive this mechanism reaction to take place?

DR. HAYES: I believe that it is not necessary to assume that the gas penetrates the lattice to any great distance as I believe that the CO-CO_2 can

act as a carrier in the grain boundaries; that combined with the diffusion of carbon as such might account for considerable of the catalysis. I just make that suggestion. We know this is what takes place in the carbonizing process and we know they are in the malleableizing process.

There is an additional statement which I wish to make in regard to the last remarks of Dr. Hoyt. It is that the chief value of the method of thermodynamics is that it is not necessary to know which of a number of possible mechanisms is the one that is really effective in bringing about a change. For if a treatment of any one of these possible mechanisms can be made, the result of this one mechanism must be the same as that of any other of the possible mechanisms. Thus the purpose that has been served by the treatment of the mechanism suggested in the paper is that it has enabled us to predict the results of the application of this gas mixture under the conditions, first that iron carbide is stable and second that it is metastable. From the results obtained the conclusion is reached that iron carbide is metastable.

It was not the purpose of this paper to establish the mechanism that is most effective in catalyzing the graphitization. What was attempted was to arrive at conclusions in regard to metastability of the carbide.

The writer does not claim to know what the mechanism is. It may be that the metal has the property of condensing small amounts of these gases or of the products of their dissociation within it in some such manner as palladium acts upon hydrogen gas, or it may be that an unstable compound such as iron carbonyl forms and moves through the metal; a cycle of chemical reactions may occur between these gases and the constituents of the metal carrying the carbon and oxygen of these gases through the metal, or it may be any other mechanism whatever.

What thermodynamics enable us to say is that if a gas at equilibrium develops at any point in such a system, the composition of the gas phase must be such as to conform to the partial pressures required by the mechanism treated in this paper. Reference to Fig. 5 will show that at 900 degrees Cent. there should be 16 per cent of CO_2 and at 1000 degrees Cent. 5 per cent at 5 atmospheres of this gas mixture at both of these temperatures.

In regard to the observations which Mr. Schwartz has given in regard to wash metal absorbing up to 5 per cent combined carbon, the statement may be made that chromium is known to be a powerful stabilizer of iron carbide. Before any great amount of effort is made at explaining this unexpected behavior it would be well to verify the results in an alloy whose composition is known. However the writer will venture the statement that if carbon migrates from carbon to this alloy in the presence of a gas phase of CO and CO_2 that equilibrium studies of the gas phase in contact with the carbon in one case and with the wash metal containing less than 5 per cent carbon in the other will show values for the gas compositions which will make possible the existence of a pressure gradient of CO from carbon to wash metal and CO_2 in the opposite direction.

The only claims made here are that the iron carbide in the alloys with which we worked is metastable.

HIGH SPEED STEELS

By F. C. A. H. Lantsberry

Abstract

The author sets forth and describes in this paper the technique of various methods of producing high speed steels in England, America and Germany. He particularly describes the methods and processes used in Sheffield, England, in the making, composition, working, hardening, tempering, and theory of high speed steels.

The writer is of the opinion that the modern tendency is to produce high speed steels in the electric furnace where refining operations make possible the use of inferior raw materials and results in greater mass production, but that the crucible process is favored in his works in Sheffield because of the inherent quality of the steel which results in a more uniform product in small quantities.

Comparative analyses of English, American and German high speed steels are shown in a table. In England high speed steels are classified according to their tungsten content. In Germany the tendency is to increase the chromium content of this commodity in comparison with English steels.

Attention is also called to the fact that on account of possessing four different types of hardness, namely: intrinsic hardness, carbon hardness, red hardness, and secondary hardness (all independent of each other), high speed steel is an interesting product, for it can, on account of its adaptability, be put to multifarious uses.

TO the production engineer, high speed steels are of enormous importance, for it is only with their aid that he can maintain the high rates of production which are essential features of modern engineering shops. High speed steels were gradually developed from the straight carbon steels, and their birth really took place when the experiments of Mushet culminated in 1860 in the production of a self-hardening steel. This material had little in common with the modern high speed steel, and was, in fact, more of a curiosity than one of use. It was in reality more of the nature of a tungsten cast iron than of a steel since its

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carbon content was well over 2 per cent, and the real interest in the material was centered in the fact that after heating to a red heat, quenching was not essential to effect hardness. Its cutting properties were not materially better than those of the straight carbon steels. In 1900 Taylor and White discovered the property of red hardness of the tungsten alloys. Since that time the further improvements in high speed steels have been very gradual, and have been directed toward the ability to cut harder materials rather than to increasing cutting speeds, although there has been a distinct improvement in the capacity for removing material.

Manufacture of High Speed Steel

High speed steels are usually made in crucible and electric furnaces, the operations of both are so well known that their description is unnecessary in this paper. The modern tendency is to produce high speed steel in electric furnaces, but in the author's works in Sheffield, England, the practice is to adhere to the old crucible process, particularly under present-day conditions when orders are small. In passing, it may be remarked, that the present-day tendency to order in small quantities is not altogether due to trade conditions, but is largely brought about by a determination on the part of steel users to operate with smaller stocks so that to the steel maker, and particularly the tool steel maker, there is a great potential danger of the tendency becoming a permanency even when conditions in the tool steel trade have again become normal.

Undoubtedly, the costs of crucible melting are appreciably higher than those of electric furnace melting; the actual costs will, of course, vary for different districts, but their ratio will be of the order of 3 to 2. There must, therefore, be reasons, other than those of the ability to produce in small quantities, which have determined the adherence to the old crucible process. In the electric furnace, excellent opportunities are afforded for refining operations so that cheaper and inferior raw materials can be used. In fact, the electric furnace could only get a foothold in the tool steel industry because of the difficulty of obtaining raw materials of the necessary purity for the crucible process during the war period. The crucible process demands the use of raw materials of the highest degree of purity so that the re-

sulting product has of necessity certain inherent qualities which are not to be found in the electric furnace product.

Again, since the greater number of the electric furnaces in use are of the arc type, the operations can only be carried out under conditions which are extremely difficult of control and occasional overheating is unavoidable. These difficulties are not met in the crucible process which, over a long period of time, undoubtedly gives a much more uniform product than the electric furnace.

To the steel user, therefore, an account of the type of "mix" used and a description of the technique of melting high speed steel in crucibles will be of considerable interest. A typical original charge for making an 18 per cent tungsten high speed steel is as follows:—

28¼	pounds	.....Swedish Bar Ends
9	pounds	.....Small Swedish Scrap
9¼	pounds	.....Washed Metal
11½	pounds	.....Tungsten Powder
3	pounds	.....Chromium Metal
30½	ounces	.....Ferrovanadium (35 per cent).

Naturally, after the manufacture has gotten into its stride, scrap is produced and this is used to replace the small Swedish scrap in subsequent mixes.

These materials are charged simultaneously through a long sheet steel hopper into the pot already heated in the furnace, and melted down in precisely the same manner as used for carbon steels. The melting operation takes from 4 to 4½ hours. Fifteen minutes before melting is complete, from 1 to 3 ounces of ferromanganese and about 2 ounces of ferrosilicon are thrown into the pot. When melting is complete, the pots are lifted from the furnace and a pill of aluminum is added to each pot. They are then teemed singly or the contents of a predetermined number of pots are combined according to the size of the ingot required. Since the surface condition of the ingot is of enormous importance in the subsequent mechanical operations, special precautions are always taken to insure that the inner surfaces of the mold are smooth and clean.

Sheffield practice has invariably been to introduce the tungsten from tungsten powder rather than from ferrotungsten. Experience has shown that the use of ferrotungsten does not produce such reliably uniform results as the use of powder because

of the variation in the composition of ferrotungsten. These variations will not affect the process of electric melting to anything like the same extent, because of the refining operations effected by this process. In the crucible melting process it is not so much that tungsten powder makes better high speed steel, but that it gives a much more uniform product over a long period of time. The melting losses are 10 to 12 per cent of the tungsten, and 30 to 40 per cent of the vanadium added, while the loss of chromium is practically nil.

Composition of High Speed Steel

In England, high speed steels are usually classified according to their tungsten content as:—

14 per cent Tungsten Steel and
18 per cent Tungsten Steel

Both grades have their specific uses, the 14 per cent class finding use in tools in which the ability to resist stresses is at least of equal importance with cutting properties such as, for example, twist drills and also for tools such as reamers and dies in which the requirements are resistance to wear rather than utilization of the property of "red hardness". The 18 per cent tungsten class certainly possess the property of cutting under more strenuous conditions, but are more brittle and should only be used where the cutting edge can be properly backed up.

The average compositions of the two classes are as follows:—

	14 per cent Class	18 per cent Class
Carbon	0.6 to 0.65	0.6 to 0.65
Tungsten	13.0 to 14.0	16.5 to 18.0
Chromium	3.0 to 4.0	3.0 to 4.0
Vanadium	0.25 to 0.5	2.25 to 0.5

During the last few years a third class of high speed steel has been introduced into this country, and is generally known as the super class. The advantage of these steels is not so much that they will cut at higher speeds, but that they will remove more metal per grind and are capable of cutting much harder material. They are made by:—

1. Increasing the vanadium content of the 18 per cent tungsten class to about 2 per cent.
2. The addition of 1 to 2 per cent molybdenum to an 18 per cent tungsten steel.

3. The addition of 5 per cent cobalt to the 18 per cent tungsten steel, or by a combination of the three.

At the same time it is customary to increase the carbon content to 0.70 to 0.75 in the super steels.

Typical analyses of these super steels are as follows:—

	per cent	per cent
Carbon	0.71	0.77
Tungsten	17.92	19.03
Chromium	3.82	3.82
Vanadium	1.26	0.64
Molybdenum		0.75
Cobalt		5.43

The principal use of these steels is in the turning of sorbitic, worn and brake-hardened railroad wheel tires. In America the 14 per cent tungsten high speed steels find very little favor, use being confined almost exclusively to 18 per cent tungsten steels. To the author this has always occasioned great surprise, because there can be no doubt that for many classes of tools 14 per cent tungsten steels are superior to 18 per cent tungsten steels, and the author has in his possession a $\frac{15}{16}$-inch twist drill which is capable of drilling holes in a 0.40 per cent carbon steel at a penetration equal to 26 inches per minute. Added to this superiority for certain classes of work, there is the fact that the costs of production of a 14 per cent tungsten steel are about 20 per cent lower than those for 18 per cent tungsten steel. Generally, there is a great similarity between the compositions of American and English 18 per cent tungsten high speed steels, the only real point of difference being that in the American steels the carbon content is 10 to 15 points higher than in the English steels.

On the continent of Europe, and particularly in Germany, the tendency is to increase the chromium content in comparison with English steels.

Typical analyses of German steels:—

	Per Cent			
Carbon	0.5	0.90	0.7	0.64
Chromium	5.0	5.1	5.0	4.3
Tungsten	18.0	18.0	14.3	16.0
Vanadium	0.15	1.2	0.5	1.12
Molybdenum	1.6	0.6	..	0.70
Cobalt	..	..	..	5.19

The following table shows comparative analyses of English, American and German high speed steels:—

	ENGLISH		AMERICAN					GERMAN			
	14% Tungsten	18% Tungsten	Super	Super							
C	0.6/0.65	0.6/0.65	0.7	0.8	0.68	0.70	0.88	0.5	0.9	0.7	0.64
Cr	3.5/4.0	3/4	3/4	3.8	4.0	3.6	4.3	5.0	5.1	5.0	4.3
W	13/14	16.5/18	18	19	16.2	17.0	18.8	18	18	14.3	16.0
V	0.2/0.5	0.25/0.5	1.26	0.6	1.3	0.83	1.3	0.15	1.2	0.5	1.12
Mo	..	..	...	0.75	1.1	...	...	1.6	0.6	...	0.7
Co	..	..	...	5.43	...	...	2.92	...	...	...	5.19

Working of High Speed Steel

The mechanical operation of converting high speed steel ingots into bars is one which must be carried out with extreme care, partly on account of the tendency of the material to air-harden and produce minute surface cracks, and partly because of the peculiar structure of the material after solidification. The microstructure of high speed steel in the cast condition is illustrated in Fig. 1. It is seen to consist of grains of decomposed complex austenite surrounded by a hard brittle eutectic consisting of a complex austenite and tungstide of iron Fe_2W. It was previously considered that the hard brittle constituent was a tungsten carbide or a double carbide of chromium and tungsten, but recent work of the Japanese school of investigators ably led by Dr. Kotaro Honda, has proven fairly definitely that at least in steels containing over 12 per cent tungsten the constituent is tungstide of iron.

In the working up of high speed steel, it is absolutely essential that this material should be broken up and evenly distributed throughout the mass, and this can only be effected by a combination of thermal treatment and ample mechanical working.

If possible, ingots of high speed steel should never be allowed to cool down between casting and the preliminary operation of forging. Ingot stripping should take place when the temperature of the outside of the ingot has fallen to about 1290 degrees Fahr. (700 degrees Cent.), thus providing ample time for complete solidification and the ingot transferred immediately to the reheating furnace where it is slowly raised to a temperature of 2190 to 2280 degrees Fahr. (1200 to 1250 degrees Cent.), after

which it is reduced to a cogged bar under a hammer. All work should be completed at 1830 degrees Fahr (1000 degrees Cent.). The resulting cogged bar is annealed at 1380 to 1470 degrees Fahr. (750 to 800 degrees Cent.) and examined for surface defects. These defects must be carefully ground out before the material is subjected to any further treatment. It will be ap-

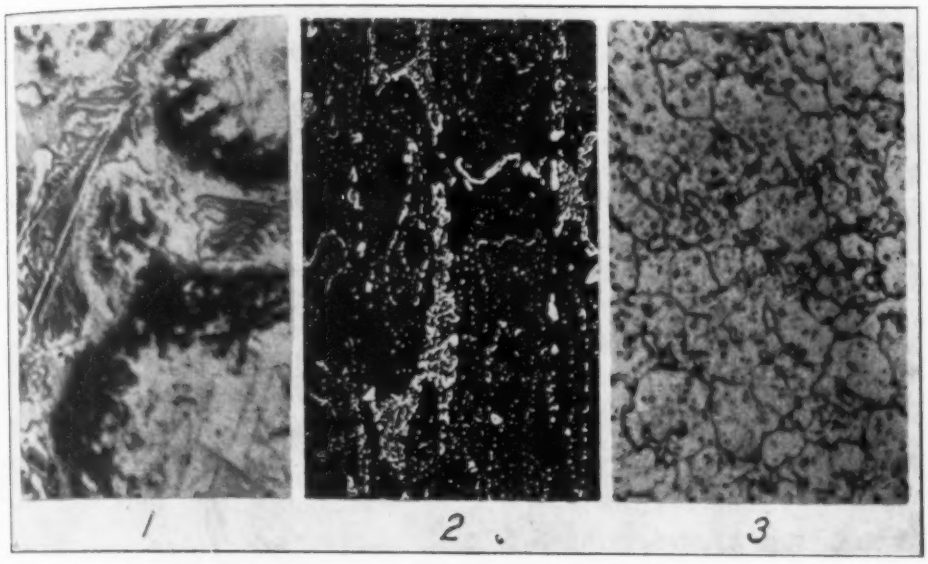


Fig. 1—Photomicrograph Showing Ingot Structure in High Speed Steel. Fig. 2—Photomicrograph of High Speed Steel Insufficiently Worked. Fig. 3—Photomicrograph of High Speed Steel Properly Hardened.

preciated that even the slightest surface defect will give rise to a great deal of trouble in the finished bar. In order to effectively dissipate the tungstide envelopes, the reduction of sectional area required is in the neighborhood of 90 per cent. Repeated trials have been made to remove the envelopes by annealing. Owing, however, to the presence of large quantities of elements in the iron, the rate at which the envelopes are dissolved in the matrix is so slow as to render the process commercially impracticable. Prolonged annealing even at temperatures as high as 2280 degrees Fahr. (1250 degrees Cent.) has little effect on the magnitude or position of the envelopes. This point is of immense practical importance to the user of high speed steel because the presence of the envelopes may lead to trouble in hardening. And yet high speed steel users frequently order bars of 6 and 8 inches in diameter for the purpose of making cutters and then blame the steel maker when the cutter ultimately cracks in the harden-

ing or during use. For the dissipation of the envelopes longitudinal extension is not essential, in fact, the "kneading" brought about by forging blanks is much more effective.

The microstructure of underworked high speed steel (actually taken from a cutter which cracked in the hardening operation) is shown in Fig. 2, from which it can readily be seen how the extended envelopes form planes of weakness in the steel.

Hardening and Tempering of High Speed Steel

High speed steel is an interesting product because it possesses four different types of hardness:—

(a) The "intrinsic" hardness as a result of its content of alloying metals
(b) Carbon hardness as a result of quenching from low temperatures not exceeding 1740 degrees Fahr. (950 degrees Cent.)
(c) Red hardness as a result of quenching from very high temperatures
(d) Secondary hardness resulting from the suitable tempering of samples in which the red hardness has been developed.

In the multifarious uses to which high speed steel is put in the various aspects of modern engineering production, advantage is taken of all these types of hardness, all of which are independent of each other. The "intrinsic" hardness will be discussed in a later section of the paper dealing with the effects of the various alloying elements and possible methods of further improving high speed steels.

Reference has already been made in an earlier part of the paper to the use of high speed steels for such tools as dies in which resistance to wear is the desideratum. Such tools are best treated by heating to 1740 degrees Fahr. (950 degrees Cent.), and allowing to cool down in the air. In this way advantage is taken of the carbon-hardening property, there being plenty of alloying elements present to render the steel air-hardening. Long experience has proved to the author that no advantage can be gained by using an 18 per cent tungsten steel rather than a 14 per cent tungsten steel. Tools hardened in this way, however, have no advantage over carbon steels so far as cutting properties

are concerned; in fact, they are not so efficient because of the lower carbon content.

High speed steel hardened at temperatures up to 1740 degrees Fahr. (950 degrees Cent.) behave like carbon steels on tempering, i. e., the hardened steel begins to soften immediately and complete softening can be effected at 750 degrees Fahr. (400 degrees Cent.). Naturally, the process of tempering is slower because of the retarding influence of the alloying elements.

To develop the property of red hardness which is essential in cutting tools, it is necessary to heat the tool to above 2100 degrees Fahr. (1150 degrees Cent.), the exact temperature required being almost independent of the tungsten content, but dependent on the carbon content. For a steel containing 0.6 per cent carbon, the optimum hardening temperature is 2280 degrees Fahr. (1250 degrees Cent.), while for a steel containing 0.75 per cent carbon the optimum hardening temperature is 2100 degrees Fahr. (1150 degrees Cent.).

In the case of tools of complicated shape, it is desirable to pre-heat the tools to 1470 degrees Fahr. (800 degrees Cent.) in order to prevent cracking by the sudden insertion into a furnace at the high temperature. Immediately after the tool has attained the high heat it should be removed from the furnace and then cooled down in still air, an air blast or in oil. The method of cooling has little influence on the properties of the finished tool unless its mass is large.

All high speed tools should, after hardening, be tempered at 1020 to 1110 degrees Fahr. (550 to 600 degrees Cent.), at which temperature the maximum secondary hardening is developed. Some workers prefer to allow the tool to develop its own secondary hardness under the heat of cutting, but there is no valid reason why the tool should not be put into a condition of maximum cutting efficiency before it is put to work, particularly in view of the fact that tempering plays an important part in removal of stresses which are liable to cause cracking, more particularly in complicated, and consequently, expensive tools.

The explanations of red hardness and secondary hardening are dealt with in the section on the theory of high speed steel.

Theory of High Speed Steels

Edwards (Journal, Iron and Steel Institute, 1908) working

on two series of tungsten-chromium steels, one with constant chromium and increasing tungsten, and the other with constant tungsten and increasing chromium, showed that so long as the initial heating temperature does not exceed 1740 degrees Fahr. (950 degrees Cent.), the addition of tungsten has no effect on the position of the critical points. If, however, heating be carried to 2400 degrees Fahr. (1320 degrees Cent.), then the subsequently obtained cooling curves show that the Ar point is split into two parts, the lower arrest occurring at 790 to 880 degrees Fahr. (420 to 470 degrees Cent.). This effect is undoubtedly caused by the tungsten present because tungstenless alloys are not affected by the temperature of initial heating. But since under normal conditions of heating or cooling chromium raises the Ar point, and tungsten does not affect it in any way, it is a logical conclusion that to whatever the special properties of high speed steel are due, they are not caused by the lowering of the critical points to below the ordinary temperature. Edwards considered it more likely that the lowered point is independent of the Ar, and is brought about by a change in a tungsten carbide, which is slowly formed at temperatures about 2190 degrees Fahr. (1200 degrees Cent.). In a tungsten-chromium steel this lowered point is found to disappear entirely when the chromium content exceeds 3 per cent, so that it would appear that in high speed steel the effect of heating to temperatures in the neighborhood of 2280 degrees Fahr. (1250 degrees Cent.) is to produce a double carbide of tungsten and chromium which forms a solid solution in gamma iron. This double carbide is more stable than the carbides of iron and tungsten, and at a higher temperature it is thrown out of solution on reheating, thus enabling the steel to continue cutting at a much higher temperature than is usual for carbon tool steels. But at that time Edwards neglected to take into account the probable function of tungstide of iron, whose existence had been foreshadowed by Harkort some years previously and subsequently confirmed by Swinden. The later chemical researches of Arnold and Read elicited the fact that Cr_3C_2, Cr_4C, WC and Fe_2W could exist in alloy steels containing these metals.

Incidentally, it might be mentioned, that vanadium also forms a carbide V_4C_3, but at the moment we are dealing only

with the essentials of high speed steels, and all these are contained in alloys of iron, carbon, chromium and tungsten. The addition of vanadium, molybdenum or cobalt may improve the properties of high speed steels in certain specific directions, but they do not in any way affect the actual fundamental properties of the steels.

Yatsevitch working at Harvard on a high speed steel containing carbon 0.62 per cent, chromium 3.0 per cent, tungsten 18.2 per cent and vanadium 0.95 per cent, did some useful work on the melting and solidifying temperatures of high speed steels. On heating, fusion commences at 2390 degrees Fahr. (1310 degrees Cent.) and ends at 2670 degrees Fahr (1465 degrees Cent.), while on cooling solidification commences at 2635 degrees Fahr. (1446 degrees Cent.) and is complete at 2310 degrees Fahr. (1284 degrees Cent.). The cooling curve shows an arrest point at 1860 degrees Fahr. (1015 degrees Cent.), which Yatsevitch ascribes to the solidification of the tungsten eutectic.

Since microscopic examination fails to reveal any evidence of fusion in samples of high speed steel which have been heated to below 2190 degrees Fahr. (1200 degrees Cent.), this arrest point cannot be due to solidification of an eutectic. But the work which has thrown most light on the theory of high speed steels is that of Honda and his pupils. They made a systematic study of the magnetic properties of the various ternary and quatenary combinations in different heat treated states, and correlated them with the microstructures. The study of the combined results of their work indicates that a properly worked and completely annealed high speed steel consists of tungstide of iron, chromium carbide and tungsten carbide, all existing in the free state. On heating to a temperature of 1545 degrees Fahr. (840 degrees Cent.) (which is the $Ar_{1\text{-}2\text{-}3}$ point) for this class of steel, the carbides dissolve in the ferrite to produce a complex austenite leaving the tungstide still uniformly distributed and microscopically indicated as white particles in etched specimens. At higher temperatures the tungstide begins to dissolve, but simultaneously another change occurs in the conversion of the carbide Cr_4C into Cr_3C_2 and free chromium thus:—

$$2\ Cr_4C \rightarrow Cr_3C_2 + 5\ Cr$$

At temperatures over 1830 degrees Fahr. (1000 degrees Cent.),

the affinity of carbon for chromium is greater than its affinity for tungsten resulting in the following change:—

$$3\ Cr + 2\ WC + 4\ Fe \rightarrow Cr_3C_2 + 2\ Fe_2W$$

This reaction is reversible only at very slow rates of cooling so that after heating at the hardening temperature 2100 to 2370 degrees Fahr. (1150 to 1300 degrees Cent.) (according to the composition) and cooling in the normal manner for hardening, a high speed steel consists of the chromium carbide Cr_3C_2, a little tungsten carbide WC, and most of the tungstide of iron Fe_2W, in a state of solid solution, while there is still a little free tungstide of iron remaining undissolved. The microstructure of a properly hardened high speed steel is shown in Fig. 3, the polyhedral grains being the solid solution (complex austenite) and the little brilliant white particles the free tungstide of iron.

Taking a line through the work of these various observers, we can arrive at a workable theory of the phenomena which occur in the hardening and tempering of high speed steel. In such an extremely complicated alloy it will be understood that the experimental difficulties are great and in the present state of our knowledge no theory could possibly be considered as final, but the following should form a satisfactory basis for further work.

In completely annealed high speed steel, the various elements exist as tungsten carbide, tungstide of iron Fe_2W, the chromium carbide Cr_4C, ferrite containing dissolved tungstide of iron and chromium, and also carbide of iron or cementite, for the laws of mass action indicate that there will be cementite in high speed steel regardless of its state. This conclusion is confirmed by the work of Grossmann and Bain who, by the analysis of electrochemical residues, proved that no matter what the state of the high speed steel from which the residues were obtained, the carbides always contained cementite. High speed steel is, therefore, similar to plain carbon steel except that both the ferrite and carbide are complex. At temperatures between 1545 and 1830 degrees Fahr. (840 and 1000 degrees Cent.), the complex carbide goes into solution in the ferrite, and gives rise to the ordinary carbide hardening.

At higher temperatures, more of the complex carbides are taken into solution, but simultaneously chemical changes occur resulting in the formation of the chromium carbide Cr_3C_2. This

releases chromium to dissolve and increase the intrinsic hardness of the solute. According to Honda it is the formation of this Cr_3C_2 which is responsible for the "red hardness" property. That this compound is formed cannot be doubted, but since the simultaneous presence of both chromium and tungsten are essential for the development of red hardness, it is impossible to attribute the whole of the red hardness to the formation of Cr_3C_2. The author is of the opinion that the real explanation of the red hardness property lies in the ability of Cr_3C_2 to form a double carbide with tungsten carbide, which double carbide is retained in solution up to higher temperatures.

The secondary hardening developed by tempering is caused by the breakdown of the austenite into martensite, possibly, though not necessarily, assisted by the conversion of the Cr_3C_2 into Cr_4C. To account for these changes, it is not necessary to assume the presence of more than 12.5 per cent of tungsten in the high speed steel. The effect of increasing the tungsten content is to increase the proportion of Fe_2W which is proved microscopically. This increase of the number of structurally free particles of tungstide of iron is the most rational explanation of the ability of 18 per cent tungsten steels to cut harder materials, than can be cut by lower tungsten steels. With regard to the other alloying metals which are to be found in high speed steels, the following remarks may be of interest.

Vanadium increases the breakdown temperature of high speed steel partly by forming the carbide V_4C_3 which associates with, and increases the stability of, the complex carbide described above, and possibly also by dissolving in the ferrite or solute and assisting in the retention of the complex carbide in solution up to higher temperature.

Molybdenum behaves like tungsten and cannot be regarded as having any effect beyond that of replacing tungsten. Thus, a steel containing 14 per cent tungsten and 2 per cent molybdenum is only to be regarded as the equivalent of an 18 per cent tungsten steel.

Cobalt does not form carbides and associates itself with the ferrite whose properties it may modify. A great deal of work has been carried out on cobalt high speed steels, and it is difficult to determine from the published results whether the pres-

ence of cobalt is advantageous, because all the steels which have given good results have contained fair proportions of vanadium.

With regard to possible future improvements in high speed steels, it has already been indicated that there is little probability of further increasing the cutting speeds. Actually there is little necessity for this as the modern high speed steels are capable of cutting mild steels at over 150 feet per minute. But there is a real need for a material which is capable of cutting steels in a toughened condition. An example, namely, that of sorbitic and brake-hardened tires has already been quoted, but there are many other instances where the ability to cut steels which have been toughened by heat treatment would be of assistance to engineers.

The two properties of high speed steel which determine its capacity for cutting are those of red hardness and intrinsic hardness. It is doubtful if the red hardness can be further increased because the breakdown temperature, which is a measure of the red hardness, has been increased to over 1290 degrees Fahr. (700 degrees Cent.) in some of the higher vanadium steels. There is little likelihood of this point being raised further. But unlike red hardness, the intrinsic hardness is not a function of the nature and distribution of the carbides in high speed steel, and there are still considerable hopes of effecting improvements along this avenue which is being explored by steel makers.

It does not lie within the scope of this paper to present a detailed account of that most interesting alloy stellite, but brief reference to it will throw light on the meaning of this property, intrinsic hardness. Stellite is an alloy consisting essentially of:—

40 to 55 per cent cobalt
15 to 33 per cent chromium
10 to 17 per cent tungsten
About 2 per cent carbon

with occasional amounts of iron and molybdenum.

The cutting properties of this alloy are not produced by heat treatment, but are inherent in the alloy itself. At ordinary temperatures the hardness of this material is appreciably lower than that of high speed steel, but its hardness falls with increasing temperature at a much lower rate. This is shown in Fig. 4 from which it will be seen that although at low temperatures

hardened high speed steel is harder than stellite, yet at 1160 degrees Fahr. (625 degrees Cent.) the hardness temperature curves cross, and at higher temperature stellite becomes the harder material. Further than this, the stellite recovers the whole of its hardness on cooling, while high speed steel does not, owing to

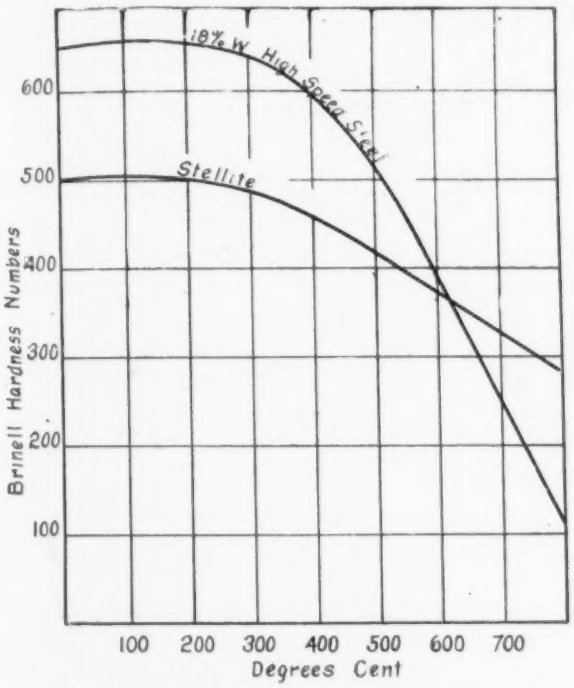


Fig. 4—Brinell Hardness Curve of a 18 Per Cent Tungsten High Speed Steel Compared with the Brinell Hardness Curve of Stellite.

the tempering effect. Stellite is a typical example of a metal whose cutting properties are inherent; another example is the diamond, and it is the property possessed by this material which the author understands as intrinsic hardness. It is to the improvements in this property on which depends further advances in high speed steels, and efforts in this direction are being made by the introduction of such metals as cobalt and molybdenum.

DISCUSSION

N. B. HOFFMAN: I feel that this contribution is certainly worthy of comment. One thing is quite certain, that the methods used in England differ a lot from the methods used in the United States, both as to composition and methods of manufacture. Mr. d'Arcambal, in the presentation of the paper, has pointed out this fact. The composition of one of the high speed steels they use, I notice contains 0.90 per cent carbon and 18 per cent tungsten. Is that correct?

A. H. D'ARCAMBAL: That is a German composition.

N. B. HOFFMAN: The Germans must have some special method of reducing the 0.90 per cent carbon ingots into the finished bars. It may be possible that they are casting this steel in very small ingot sections, but in that event their carbide reduction would suffer. High speed steel practice in this country has shown that with the higher carbons, say 0.80 per cent or over, are hard to forge or roll, due to brittleness and in this event the manufacturing losses will be much higher than in the case of lower carbon steels. It has not been demonstrated as yet that the higher carbon steels are beneficial or desirable.

Another point brought out in Mr. Lantsberry's paper is that the English are using a carbon content considerably lower than that adopted by us in this country, namely, 0.60 to 0.65 per cent. I believe that metallurgists in this country have found the standard high speed steel to take its maximum hardness at about 0.685 per cent. Is this correct?

A. H. D'ARCAMBAL: About 0.70 per cent.

N. B. HOFFMAN: In that event, you see, the English are working considerably below the point where we get our maximum hardness. The question of the superiority of steel melted in the crucible as compared to that manufactured in the electric furnace is another rather interesting point. There has been much discussion in recent years relative to this point, but to my knowledge no substantial evidence has as yet been produced, giving reliable figures substantiating the fact that one steel is superior to the other. I used to feel that high speed steel melted in the crucible was superior to that melted in the electric furnace, but at the present time, I believe that, it is an open question. I would like to know if Mr. d'Arcambal has any information on this point. Has any reliable data ever been produced to prove that electric furnace high speed steel is superior to that melted in crucibles, or vice versa?

A. H. D'ARCAMBAL: Mr. Chairman, answering Dr. Hoffman's question, we are getting better results today from tools made of electric furnace high speed steel than we ever did with crucible-melted high speed steel. Whether that is due to the fact that we know a little bit more today than we did a few years ago about hardening the tools, that might have some result. But the steels we are getting today, all high speed steels, regardless of type, that are melted in the electric furnace are very satisfactory.

You might be interested in knowing that one of our own mills in this country markets a high speed steel running 0.90 to 0.95 per cent carbon, with 18 per cent tungsten, 3¼ per cent vanadium and 4 per cent chromium. We are working with that steel now, but I am unable to state just how tools made of that high carbon, high vanadium steel will function as compared with the usual 0.70 per cent carbon, 18 per cent tungsten.

I am glad that Dr. Hoffman brought out the point about the low carbon content. We find that steels running 0.70 to 0.75 per cent carbon give much better results when used for metal cutting tools than steels with 0.60 to 0.65 per cent carbon. In our opinion there is no comparison, about two to one so far as cutting is concerned.

One of the things about Mr. Lantsberry's heat treating temperature is the hardening point. He states they harden the 0.75 per cent carbon steel

at 2100 degrees Fahr., whereas we harden such a steel at 2340 degrees Fahr. I have had letters from men in England asking us why we harden at such high temperature. The answer is, it gives better tools.

M. A. GROSSMANN: I should like to add a few comments. I am sorry that Mr. Lantsberry is not here to answer some questions that arise. The first point that he made about crucible and electric furnace melting raises quite a question. He stated first that crucible-melted steel required the use of more expensive materials, that with the electric furnace-melted steel cheaper raw materials could be used. Unless that statement is defined a little more closely, I think it must be questioned, because expensiveness is not in itself a metallurgical virtue. Certainly there are numerous processes where cheaper raw materials can be used, because they are not so useful for other processes and because they are available. Also, in comparing crucible furnace melting with electric furnace melting, the conditions are different in this country from those in England. In this country we melt in graphite crucibles on account of the need of transportation, and the graphite itself has something of a refining action during the melting. In England, while I am not at all able to discuss the clay crucible melting, the conditions are certainly different. In this country it has been found that adequately hard and useful high speed steels can be prepared in the electric furnace, and the economic conditions have certainly permitted the use of electric high speed steels almost throughout.

In comparing 18 per cent tungsten steels and 14 per cent tungsten steels, the matter of hardening temperature was also brought out very well by Mr. d'Arcambal. It is true that the higher hardening temperatures cause more of the tungsten to go into solution and therefore to be effective and useful. Additional tungsten tends to prevent grain growth at high temperatures, which makes it possible to use higher hardening temperatures and still avoid coarse grain. Additional cutting efficiency is thus obtained, while reducing the possibility of brittleness from coarse grain at the high hardening temperatures.

There was also mentioned the matter of cobalt, which, as Mr. Lantsberry says, is probably associated with the ferrite, but it seems to give a stiffer matrix at high temperatures and there seems to be little doubt from American experience that cobalt steels do give somewhat better service.

E. C. BAIN: During the reading of this paper it has seemed to me that there has been little consideration given to the opinions of such investigators as Gill, Bowman, Jeffries, Grossmann and Maurer, and for that reason I am prompted to say a few words. Anyone who finds the study of the nature of high speed steel leading him toward physical considerations should not be prejudiced against the contemplation of such chemical phenomena as are transpiring in the preparation of high speed tools—but it may be doubted that our fundamental understanding is to be improved by theories involving carbide and tungstide reactions which do not correlate themselves directly with physical manifestations.

I hope I have not misconstrued the sentences of this paper, but I have gleaned the impression that high speed steel is said to be largely austenite

and iron-tungstide particles. If so, I find the experimental evidence pointing to very different conclusions. In the first place the X-rays have shown most reliably that while austenite is present after the high quench—even then the alloy is not dominantly austenitic and in all other conditions of heat treatment it is less so. In annealed or thoroughly tempered high speed steel there is essentially no austenite. A consideration of the various properties of this steel as modified by tempering temperature as plotted by Grossmann and Bain would lead one to the same conclusion.

That the particles of hard constitutent in high speed steel are chiefly carbide seems to have been reasonably well established. Doubtless there may be some precipitation of tungstide particles, but the chief reactions seem to involve carbide. The present speaker has reported the results of the analysis of the electrolytically isolated carbides and the carbon content is well over 2.0 per cent. When a bar of high speed steel is made anode in a suitable hydrochloric acid electrolyte the solid solution phase may be dissolved out leaving a skeleton of the hard particles essentially unattacked. This residue when dried is brilliant gray in color and of metallic luster. Hence it is free from graphite. Yet it analyzes 2.3 per cent carbon. Considering that the residue contains about 60.0 per cent tungsten the hard particles may well be a carbide of reasonable formula. At any rate, the idea of a substantially tungstide material is not tenable.

It occurs to me that it might be well to mention that carefully carried out high temperature hardness measurements on high speed steel in comparison with stellite indicate that, as has been said, stellite at room temperature is slightly softer by Rockwell test than the steel, but that the curves (hardness versus temperature) cross at about 1100 degrees Fahr. (600 degrees Cent.) at which temperature both materials have a hardness of about 57 Rockwell "C". Above this temperature the high speed steel softens rapidly while the stellite retains its hardness tenaciously softening only slowly with the rise of a few hundred degrees.

CHAIRMAN JEFFRIES: I would like to call attention to one feature of the paper, and that is the theory of hardening. The theory presented by Mr. Lantsberry is what might be called the older theory, which is complex and assumes the formation of chemical compounds much as though those elements were out in the free state where they had every opportunity to get together and form these various compounds. Such a condition does not obtain in a piece of steel. These compounds, or associations, form only as the various atoms are able to get together, and inasmuch as the atoms can only get together by diffusion through a solid, it can be seen that the formation of these various compounds becomes a difficult matter.

Now, in addition to using the old formula for iron tungstide (Fe_2W), which Sykes has now well demonstrated to be Fe_3W_2, a number of other compounds are given with formulae as though they had been isolated in high speed steel and as though it were an assured thing that such compounds were present. Considering the whole subject of the theory on the simple conception held by a great many metallurgists in this country, and I am quite sure a good many metallurgists in Europe as well, we have at least a simple and

plausible basis as a beginning. The theory may not be as simple as expounded, yet it is much better to start on a simple, plausible theory and then introduce the complexities gradually than to start with so much complexity that we are baffled in the beginning. The conception of a great many people over here, and, as I say, a great many people in Europe, too, is that the various alloying elements go into solid solution in the iron at the relatively high temperatures used prior to quenching. At the high temperatures, then, we have a maximum of solution of those various elements. Then, in quenching, the cooling is so rapid that these substances do not precipitate as they would on slow cooling and there is trapped a considerable quantity of these materials in supersaturated solid solution.

On reheating a number of things happen. The first thing that can happen is that the carbon, being a very small atom and being capable of diffusing in solid iron at room temperature, and because of the fact that iron atoms are available on all sides, forms the simple iron carbide Fe_3C. That effect is strongly suggested by the work of Grossmann, Gill, and others. Then, as the temperature increases, we finally reach a point where the larger atoms, such as chromium and vanadium, and finally, tungsten can diffuse in solid iron.

Now, at about 1110 degrees Fahr. (600 degrees Cent.) the tungsten atoms are able to diffuse slowly in the iron alloy. The formation of intermetallic compounds is a matter of getting a number of atoms together. One atom of tungsten probably would not form the complex compound with carbon and iron. It takes more than one atom of tungsten and probably several atoms to form the crystalline complex carbide. In the solid solution state the tungsten atoms are as far apart as they can get and increasing the temperature from room temperature to 1110 degrees Fahr. (600 degrees Cent.) permits the tungsten atoms to diffuse and get together and therefore permits the particles of the complex carbide containing tungsten, iron, chromium and vanadium to form. The secondary hardness is perhaps partly due to formation of myriads of these fine particles of hard carbide and partly to the transformation of the balance of the austenite to the alpha iron modification. The probable reason that the fine particles do not become large is because of the sluggishness of diffusion of the tungsten. That is why high speed steel holds its red hardness.

Ordinary steel, as we know, heated to the same temperature produces a particle size of carbide Fe_3C, so large that it can be readily seen under the microscope. The minute particles of the complex tungsten carbide which form at about 1110 degrees Fahr. (600 degrees Cent.) cannot be seen with the microscope, at least, if the tempering time is relatively short, but the fact that those particles remain small at 600 degrees Cent., just the same as the particles of Fe_3C, or cementite in carbon steel remain small at 390 degrees Fahr. (200 degrees Cent.), accounts readily for the red hardness of high speed steel.

The statement of the theory given above is much in skeleton form, but I just wanted to call attention to it because it is so simple and because all of the facts which we have on high speed steel are consistent with that theory.

(Continued on Page 803)

ACTION OF PURE CARBON MONOXIDE UPON IRON AT ELEVATED TEMPERATURES

By W. P. Fishel and John F. Wooddell

Abstract

A study has been made of the action of pure carbon monoxide upon commercially pure iron in order to show the mechanism of the reaction and to investigate the ability of carbon monoxide to penetrate iron.

IT is generally believed that carbon monoxide reacts with iron according to the following reaction:

$$3Fe + 2CO \rightleftarrows Fe_3C + CO_2 + \text{heat.}$$

As far as the authors have been able to learn, this equation has not, except in one instance,[1] been investigated from the standpoint of the carbon dioxide formed. There seems to be a disagreement as to whether this reaction takes place at the surface of the iron only, and the combined carbon then diffuses toward the interior, or whether carbon monoxide is able to penetrate hot iron and react beneath the surface as well. The latter possibility would account for the greater depth of case obtained with carbon monoxide than with other cementing agents. If carbon monoxide penetrates iron and reacts beneath the surface, the carbon dioxide formed must be able to diffuse out again or the reaction would speedily come to an end.

Carbon monoxide, when heated, decomposes according to the equation:

$$2CO \rightleftarrows CO_2 + C + 39{,}000 \text{ calories}$$

The reaction is reversible and the amounts of carbon dioxide, according to Rhead and Wheeler,[2] in equilibrium with carbon monoxide and carbon at approximately one atmosphere pressure and at the following temperatures; 1470, 1650, 1740, 1830, 1920 and 2010 degrees Fahr. (800, 900, 950, 1000, 1050 and 1100

[1] Charpy, *Comptes Rendus*, 137, 120 (1903).

[2] Rhead and Wheeler, *Journal*, American Chemical Society, 97, 2179, (1910).

The authors, W. P. Fishel, assistant professor of metallurgy and J. F. Wooddell, graduate student, department of chemistry, Vanderbilt University, Nashville, Tenn.

degrees Cent.) are: 6.2, 2.2, 1.3, 0.6, 0.4 and 0.2 per cent CO_2 respectively. Iron, nickel and other metals are said to catalyze this reaction. With iron at least it is not a case of true catalysis. The action may be considered to take place by two steps,

$$(1) \quad 2CO \rightleftarrows CO_2 + C$$
$$(2) \quad 3Fe + C \rightleftarrows Fe_3C$$

Combining these we get the equation assumed above.

The authors will attempt to show that the carbon dioxide obtained in their experiments could not all have resulted from step (1), since the amounts obtained at each temperature exceeded the equilibrium values.

In order to show that carbon dioxide and iron carbide are formed by the action of carbon monoxide upon iron, and to investigate the effect of temperature upon the depth of case and amount of carbon deposited, the following experiments were carried out. Pure carbon monoxide was passed over samples of Armco iron, heated in a silica tube to various temperatures. The carbon monoxide was prepared by dropping chemically pure formic acid into warm concentrated sulphuric acid. The gas formed was passed through potassium hydroxide and stored in tanks over water. Before entering the silica tube the gas was further purified and dried by passing through a train consisting of alkaline pyrogallic acid, barium hydroxide, calcium chloride and concentrated sulphuric acid. The barium hydroxide served as a visible indicator for carbon dioxide. At no time did this solution become cloudy. The exit gas was collected over water and analyzed for carbon monoxide and carbon dioxide. Approximately two liters of gas passed through the aparatus per hour. The rate of flow varied considerably from time to time, which accounts for some of the irregularities in the composition of the exit gas. All samples were cylindrical ($1\frac{1}{2} \times \frac{1}{2}$-inch) in shape and were cut from the same bar of Armco iron. The samples were freed from scale and weighed before they were placed in the furnace. They were numbered to correspond to the runs. Thus sample 14 was introduced for the first time in run 14. The sample marked 14-15 is the larger part of 14 and has been subjected to run 14 followed by run 15.

Each run was conducted as follows: When the furnace had

reached the desired temperature, the tube was flushed out with carbon monoxide and the rate of flow adjusted. The gas was allowed to pass through the apparatus until analysis of the exit gas showed that the tube had been swept free of air and that the decomposition of the carbon monoxide had become constant for that temperature and rate of flow. Several 100-cubic centi-

Table I
Carbon Monoxide Passed Over Armco Iron

Run No.	Composition of Exit Gas Before Introduction of Samples Per Cent CO_2 #	Composition of Exit Gas After Introduction of Samples Per Cent CO_2 #	Samples Present	Per Cent Carbon added Calc. from Increase in Weight	Total Penetration in Millimeters
Run 19 1470° F. Time 10 hrs.	4.0 3.6 3.4(3)	4.2 4.0 3.8 3.6 3.4(2) 3.6(2)	19 18–19	.067 .020	0.5 6.0
Run 15 1740° F. Time 10 hrs.	0.8 0.7(3)	1.8 1.6(8)	15 14–15	.133 .039	1.0 3.5
Run 17 1830° F. Time 10 hrs.	1.3 1.3 1.3 1.3	2.4(3) 2.2(2) 2.0(3)	17 16–17	.140 .075	1.8 3.0
Run 16 1830° F. Time 10 hrs.	0.8(5)	2.6 2.2(5) 2.4(2)	16 15–16 14–15–16	.121 .078 .052	1.6 2.0 4.8
Run 14 1920° F. Time 18 hrs.	2.0(3) 1.6(2)	2.6(5) 2.4(4) 2.3(2) 2.2(4) 2.0(3)	14	.226	4.0
Run 18 2010° F. Time 10 hrs.	1.8(4)	2.6 2.4(2) 2.2(5)	18 17–18 16–17–18	.163 .030 .053	6.0 to center to center

Remainder of gas was carbon monoxide. Numbers in brackets represent the number of separate and consecutive determinations which gave that result.

meter samples were then taken over a period of one hour and analyzed for carbon monoxide and carbon dioxide. Having determined the amount of carbon dioxide in the exit gas, the weighed samples were introduced into the tube and a complete analysis was made of the exit gas each hour thereafter. Results of the gas analyses, both before and after the introduction of the samples, as well as the depth of penetration and carbon absorbed, are recorded in Table I.

At the completion of a run the samples were furnace-cooled, weighed and sectioned at a distance from one end equal to the diameter of the sample. The smaller part was examined microscopically and the larger part was put aside to be used in the next run.

The amount of carbon dioxide formed after the samples were introduced is, at all temperatures, greater than that given by Rhead and Wheeler as the equilibrium values for the system CO, CO_2 and carbon. It is also greater than the carbon dioxide formed from the decomposition of carbon monoxide before the samples were introduced. The excess carbon dioxide must result from the action of carbon monoxide upon iron.

The amount of carbon dioxide formed from the decomposition of the carbon monoxide, before the samples were introduced, decreases with increasing temperature as it should for reversible exothermic reactions, and is always less than the equilibrium value, until a temperature of 1930 degrees Fahr. (1050 degrees Cent.) is reached. At this temperature, the amount of carbon dioxide increases abruptly and exceeds the equilibrium value. At 1930 degrees Fahr. (1050 degrees Cent.) and above a dark gray scale resembling silicon carbide formed on the sides of the silica tube, indicating that the carbon monoxide had reacted with the silica tube in some manner to give abnormal values of carbon dioxide.

The photomicrographs, (Figs. 2 to 9), all of which were taken at 48 diameters, show the effect of time and temperature upon the penetration and the carbon concentration. A high temperature gives a low carbon case with deep penetration, while a low temperature produces a higher carbon case with less penetration. Photomicrograph Fig. 4 of the sample carburized for ten hours at 1830 degrees Fahr. (1000 degrees Cent.) and photomicrograph Fig. 7 of the sample carburized first for ten hours at 1830 degrees Fahr. (1000 degrees Cent.) followed by ten hours at 2010 degrees Fahr. (1100 degrees Cent.), show that the maximum per cent of carbon introduced at the lower temperature has been reduced by the treatment at the higher temperature. The total carbon introduced, however, increases with increasing temperature.

In order to test the ability of carbon monoxide and dioxide

to penetrate iron, heated to various temperatures, the apparatus shown in Fig. 1 was constructed.

Cylinder (A) (inside), made from one piece of Armco iron, by boring a $\frac{5}{16}$-inch hole through the center and turning down the outside to leave walls $\frac{3}{32}$-inch thick, is threaded into cylinder B (outside). A is connected by a ⅛-inch pipe to an automatic Toepler pump, modeled after McLaughlin and Brown.[3] This

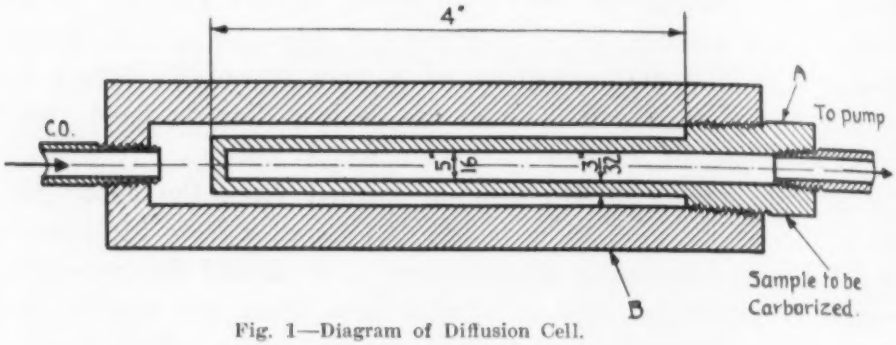


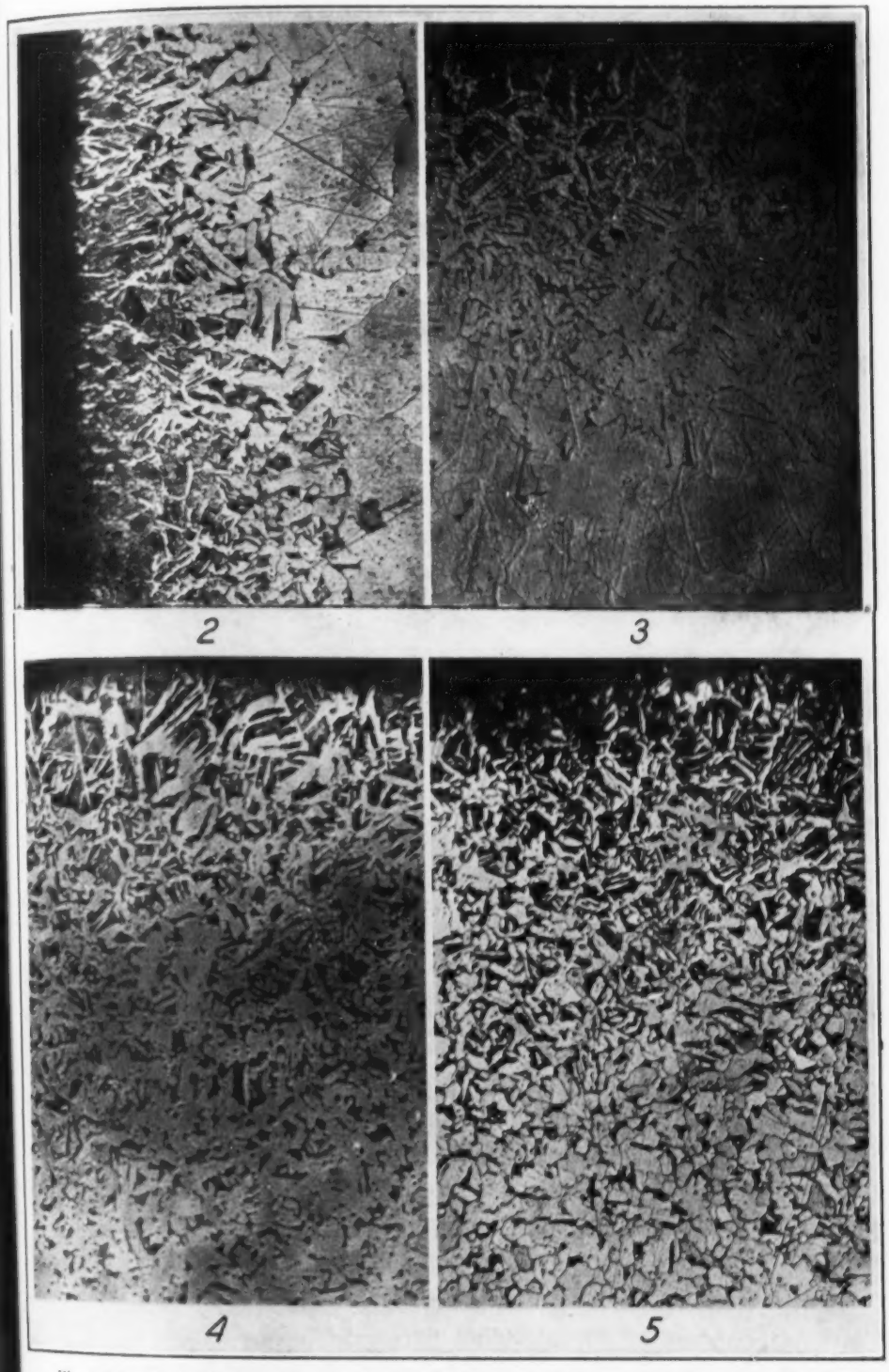
Fig. 1—Diagram of Diffusion Cell.

pump permits the exhaust gases to be collected over mercury. Pure carbon monoxide, under a pressure of ½ pound per square inch is led into B. Any gas entering A must do so by diffusing through the walls of A. The gases pumped out of A should be those formed by the action of carbon monoxide upon iron during its passage through the walls of A together with some unused carbon monoxide.

By means of a two-way stopcock, B was evacuated, filled with carbon monoxide and again evacuated. This was repeated until all air was removed from B. The stopcock was then turned connecting B with the carbon monoxide tank. A having been connected to the Toepler pump and evacuated, the cell was brought to temperature, and after about three hours had elapsed, the gas pumped out of A was collected and analyzed. The Toepler pump was kept running during the entire time of heating and maintained an average vacuum of 4 millimeters. The analyses of the gas collected are recorded in Table II.

At 1740 degrees Fahr. (950 degrees Cent.) about 5 hours were required to collect 30 cubic centimeters of gas, which contained approximately 1 cubic centimeter of carbon dioxide and 4 cubic centimeters of carbon monoxide. During the forty hours

[3]McLaughlin and Brown, *Journal*, American Chemical Society, Vol. 47, 613.



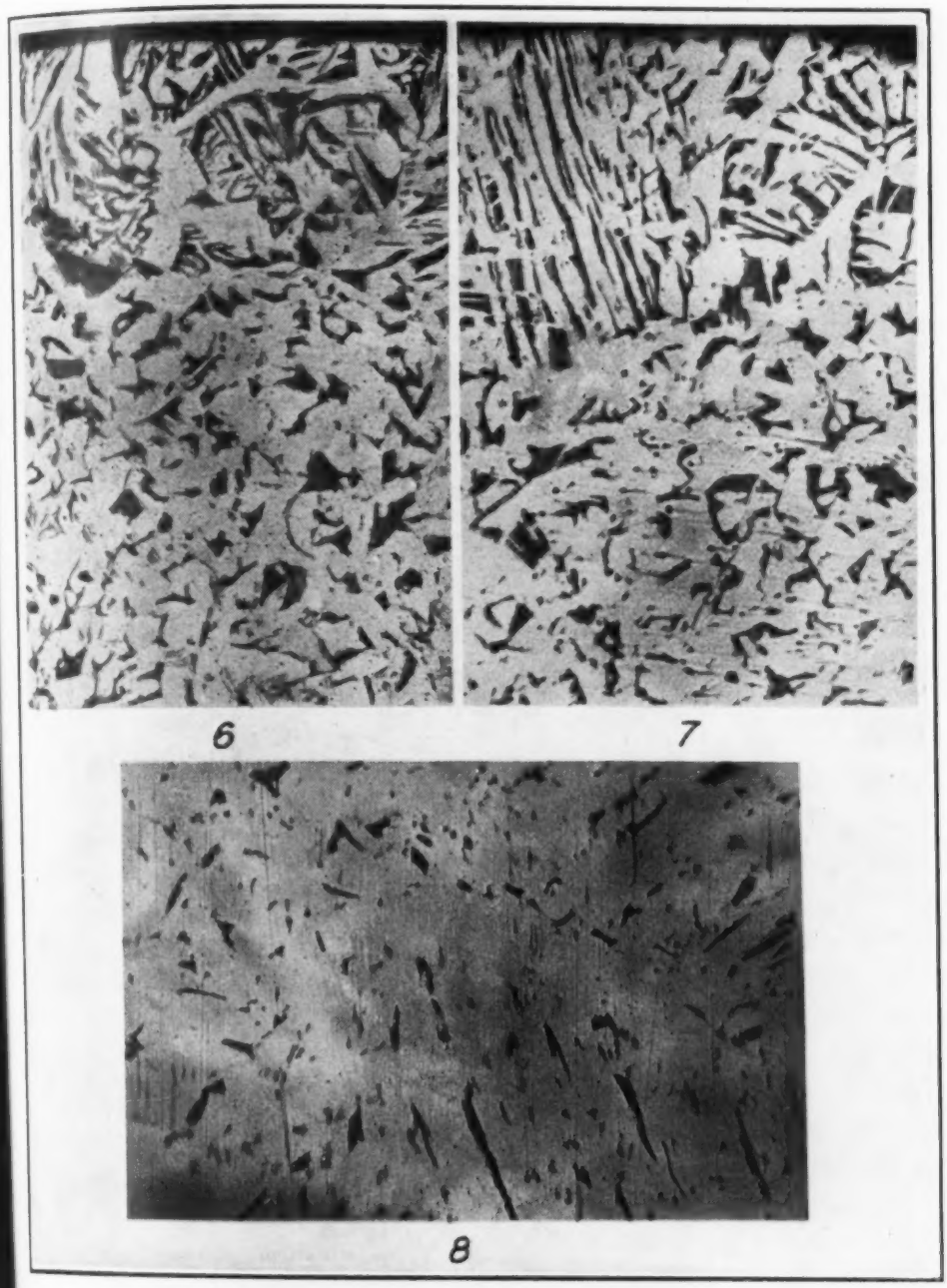
Figs. 2-5—Photomicrographs Showing the Effect of Time and Temperature Upon the Penetration and the Carbon Concentration in Armco Iron. Mag. 48 X. Fig. 2—Run 19, Carburized 10 Hours at 1470 degrees Fahr. with a Penetration of 0.5 millimeters. Fig. 3—Run 15, Carburized 10 Hours at 1740 degrees Fahr. with a Penetration of 1 millimeter. Fig. 4—Run 17, Carburized 10 Hours at 1830 degrees Fahr. with a Penetration of 1.8 millimeters. Fig. 5—Run 16, Carburized 10 Hours at 1830 degrees Fahr. with a Penetration of 1.6 millimeters.

at 1740 degrees Fahr. (950 degrees Cent.), approximately 9 cubic centimeters of CO_2 and 34 cubic centimeters of CO were pulled through the walls of cylinder A. This should be a measure of the ability of the gases to penetrate iron under the conditions of the experiment. At 2010 degrees Fahr. (1100 degrees Cent.)

Table II
Diffusion Data

Cubic Centimeters of Gas Analyzed	CO_2	CO	O_2	N_2	CO/CO_2	Temperature Deg. C.	Temperature Deg. F.
			COMPOSITION OF EXHAUST GASES IN % BY VOLUME				
19.0	4.2	22.1	12.6	69.1	5.0	900	1650
22.4	5.4	17.0	11.7	65.9	3.2	950	1740
23.2	4.3	14.2	11.6	69.8	3.3	950	1740
27.7	4.0	12.3	11.5	72.2	3.1	950	1740
29.7	2.1	6.8	11.2	79.9	3.3	950	1740
17.6	3.4	18.2	13.8	64.6	5.3	950	1740
33.6	4.2	23.8	12.6	59.4	3.0	950	1740
31.4	3.8	19.1	8.3	68.8	5.0	950	1740
31.6	3.8	12.3	12.6	71.3	3.2	950	1740
Total gas 233 cubic centimeters	9.1	34	Total time at 1740 deg. Fahr. (950 deg. Cent.) 40 hours.				
31.4	4.6	20.4	8.9	66.2	4.6	1100	2010
30.6	5.8	24.2	9.1	60.9	4.1	1050	1920
36.1	1.9	21.1	8.7	68.3	4.5	1100	2010
36.4	3.8	20.3	9.6	66.2	5.0	1100	2010
28.2	4.3	16.3	16.4	63.0	4.0	1100	2010
36.1	5.5	21.4	12.2	60.9	3.6	1100	2010
36.0	3.6	19.7	10.0	66.7	5.4	1100	2010
30.7	5.5	23.4	10.4	60.6	4.3	1100	2010
31.6	5.1	27.2	9.5	58.2	5.4	1090	1995
32.2	4.4	30.7	9.3	55.9	7.0	1050	1920
Total gas 900 cubic centimeters	45.0	207.0	Total time at 2010 deg. Fahr. (1100 deg. Cent.) 80 hours.				

the penetration rates, as estimated by the time required to collect 30 cubic centimeters of exhaust gas, were about twice as great. At 2010 degrees Fahr. (1100 degrees Cent.) some three hours were required to collect 30 cubic centimeters of the gas, which contained 5 per cent CO_2 and 23 per cent CO. During the eighty hours at this temperature approximately 45 cubic centimeters CO_2 and 207 cubic centimeters CO were collected. From Table II it will be seen that the rate of diffusion of both gases increased with increasing temperature, but not in equal proportion. The ratio of CO to CO_2 in the exhaust gases increased with increasing temperature as it should for an exothermic reaction. The composition of the gas just after diffusing through the walls of A, should be near the equilibrium values for the system Fe_3C CO and CO_2.



Figs. 6-8—Photomicrographs Showing the Effect of Time and Temperature Upon the Penetration and Carbon Concentration in Armco Iron. Mag. 48 X. Fig. 6—Run 18, Carburized 10 Hours at 2010 degrees Fahr. with a Penetration of 6 millimeters. Top of Photomicrograph is Edge of Sample. Fig. 7—Runs 17 and 18, Carburized 10 Hours at 1830 degrees Fahr. Followed by 10 Hours at 2010 degrees Fahr. Top of Photomicrograph is Edge of Sample. Fig. 8—Center of Sample 17-18. Compare Figs. 5, 6 and 7 for Carbon Content.

Although all precautions were taken to prevent the introduction of air, and both pump and cylinder were tested prior

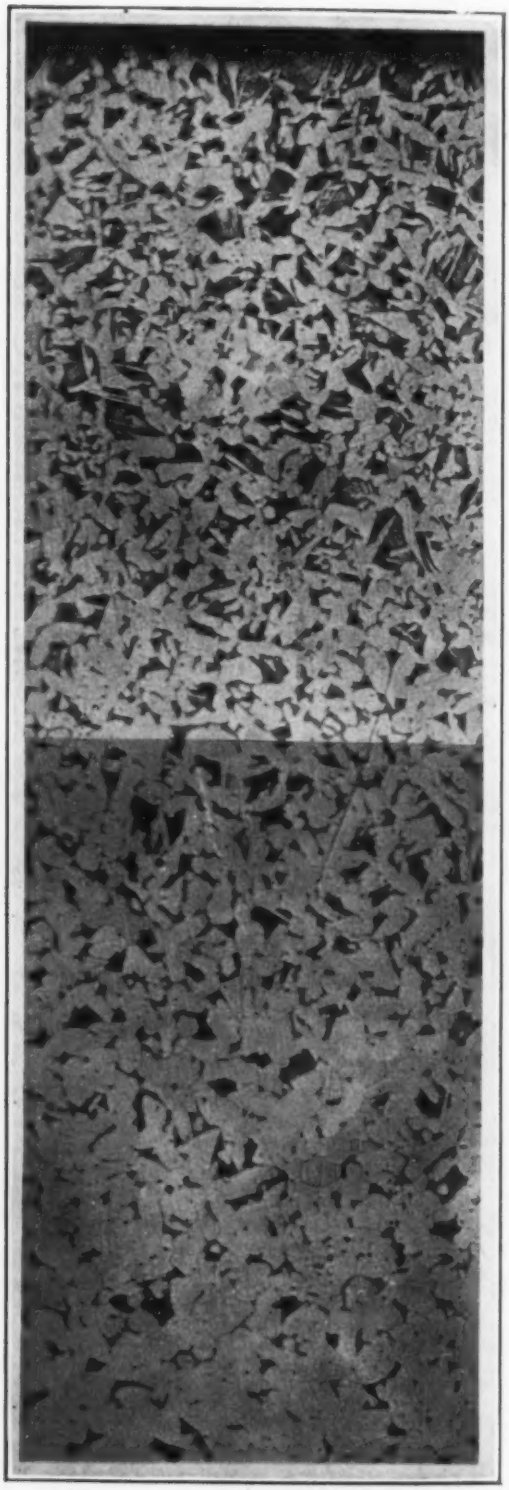


Fig. 9—Photomicrograph Showing the Effect of Time and Temperature Upon the Penetration and Carbon Concentration in Armco Iron. Mag. 48 X. From Edge (Top) to Center of Run 14. Carburized 18 Hours at 1920 degrees Fahr. with a Penetration of 4 millimeters.

to use, the presence of nitrogen and oxygen in the exhaust gas indicate small leaks. However, the proportion of nitrogen to oxygen is greater than that of nitrogen to oxygen in the air. The question arises whether the carbon dioxide found might not have resulted from the oxidation of carbon monoxide, after it had entered A, by the oxygen of the air which had leaked into the train or pump.

In all cases except one, the oxygen plus half the carbon dioxide is less (calculated to air basis) than is required by the nitrogen present. Analysis showed both carbon monoxide and oxygen were left, and this together with the fact that the ratio of CO to CO_2 changes with temperature, indicate that no oxygen entered the apparatus at a point where the temperature was great enough to cause it to unite with carbon monoxide. The excess nitrogen (over that found in air) may be accounted for by diffusion through the hot walls of that portion of the delivery pipe which fits into cylinder A. Any oxygen attempting to diffuse would react with the iron and thus be prevented, while nitrogen could pass through. Excessive oxidation was noticed at this point.

At the completion of the 120-hour run, cylinder A was furnace cooled, sectioned and examined microscopically for carbon. No pearlite was visible. This is due to the fact that the carbon monoxide in cylinder B reacted with the iron in contact with it and soon reached equilibrium. The reaction must then cease until the gaseous product of the reaction (CO_2) is removed. Such removal is brought about only by the carbon dioxide diffusing through the walls of cylinder A. Since only approximately 54 cubic centimeters of CO_2 were formed, and since no carburization can take place without the formation of CO_2, one would not expect the cylinder to be carburized to an extent that could be detected by the microscope. These conditions prevented the surface reaction of carbon monoxide with iron and permitted its penetration and subsurface action to be studied.

Summary

The following conclusions may be drawn from the data presented.

1. Pure carbon monoxide reacts with iron at temperatures

between 1470 and 2010 degrees Fahr. (800 and 1100 degrees Cent.) to form iron carbide and carbon dioxide, according to the equation:

$$3Fe + 2CO \rightleftarrows Fe_3C + CO_2 + \text{heat.}$$

2. The reaction is reversible and obeys the laws of reversible reactions.

3. The depth of case increases with increase of temperature, while the maximum carbon content in the case decreases. Total carbon introduced increases with increasing temperature.

4. Both carbon monoxide and dioxide penetrate iron very slowly at 1740 and 2010 degrees Fahr (950 and 1100 degrees Cent.). This rate would not account for the depth of case produced by carbon monoxide.

MECHANICAL AND MACHINING PROPERTIES OF AN ANNEALED CAST IRON

By G. C. Priester and F. J. Curran

Abstract

The authors in this paper set forth the results of their study of the machining and mechanical properties of a cast iron used in the manufacture of pistons when subjected to various annealing temperatures. The machining properties of the annealed pistons were studied in a modern production plant and the various processes are described. The results of heat treatment, the chemical analysis, machining properties, transverse tests, tensile and hardness tests, compressive and shear tests and the effect of time on annealing are tabulated. The findings of a study of the microstructure of the test bars are shown in photomicrographs. A brief summary sets forth the principal results of the investigation.

CAST iron differs materially from the other iron-carbon series of alloys, not only in chemical constitution but also in physical characteristics. Because of its variable constitution, a slight change in the manufacturing process will cause a wide variation in its physical properties. On this account, the foundryman should know the use to which his castings are applied in order to properly control their manufacture and to insure their possessing the requisite properties. Therefore, in order to make a thorough study of the properties of cast iron, it is necessary to know the complete history of the product.

Ordinary gray cast iron is seldom heat treated solely to improve its mechanical properties, probably because the cost of heat treatment often greatly exceeds the little increase in properties. However, when other properties are taken into consideration, it is often more economical to make use of a heat treatment. In the case of piston castings, it is extremely desirable for the material to have excellent machining properties because the cost of casting the piston is small in comparison to the machining cost of the finished product. While gray cast iron can be machined, its machining properties can be improved to a remarkable degree by proper

A paper presented before the eighth annual convention of the Society, Chicago, September 20 to 24, 1926. Of the authors George C. Priester is associate professor of mathematics and mechanics, University of Minnesota, Minneapolis, and Francis J. Curran was formerly metallurgist, Arrow Head Steel Products Co., Minneapolis.

heat treatment. Then, too, an unannealed piston casting will show distortion after it has been machined, because of the variable internal casting stresses. To eliminate this distortion it is necessary either to age the casting for several months after the rough machining or to anneal it at a temperature which will relieve these stresses. In the aging process it is necessary to tie up considerable capital in unfinished castings; but if the annealing process is used, castings can be machined immediately into a marketable product.

A study was therefore made of the machining and mechanical properties of a cast iron used in the manufacture of pistons when subjected to various annealing temperatures. Since hardness is a factor in determining the ease with which cast iron may be machined, it was decided first, to study cast iron which was suitable for piston material and to heat treat it in order that it be as soft as possible; then to study the machining and mechanical properties of the iron. To reduce the hardness of cast iron it must be heated to a temperature varying from 500 to 1600 degrees Fahr., depending upon the hardness desired.

This paper presents only a study made on this material when it was heated above the critical temperature and allowed to cool partially in the furnace. A commercial heat treating furnace was charged with its average load of pistons and A. S. T. M. arbitration test bars in sufficient numbers to carry out a study of the mechanical properties of this cast iron. It is to be noted that the test bars received exactly the same treatment as the pistons. This would tend to eliminate any difference in mechanical properties of the test bars and commercial pistons due to heat treatment in a laboratory furnace with its small charge and rapid heating capacity.

After the above charge had been annealed, the machining properties of the pistons were studied, and the arbitration bars were tested for their properties. In order that a comparison could be made the pistons and arbitration bars were tested in the unannealed condition. And hardness tests were made on pistons used by the manufacturers of automobiles. This test was carried out on new and used pistons in order to determine, if possible, the effect on their hardness after they had been subjected to actual commercial service, and to determine what hardness of pistons was commonly used by automobile manufacturers.

The machining properties of the annealed pistons were studied in a modern production plant. The pistons were routed through

the plant in the same manner as the stock product. The speed of cutting, the feed and depth of cut, and the breakdown of the cutting tool were determined for the annealed and unannealed castings. The mechanical properties of the A. S. T. M. bars were determined first on an autographic transverse testing machine. The ends were then machined to specimens suitable for tensile, compressive, shear, and hardness tests. The tensile, compressive, and shear tests were performed on a universal three-screw testing machine. Hardness was determined on a scleroscope and a Brinell machine.

The tensile test specimen was similar to the standard tensile specimen for gray iron specified in the A. S. S. T. Handbook. The difference consisted in using a 0.750-inch diameter and a 2-inch gage length in order that a suitable extensometer might be attached to it. The compression specimen, which was round, was 0.750 inches in diameter and 3 inches long when used with the extensometer; but in order to fracture the specimen, it was necessary to reduce its length to twice the diameter. An attempt was made to use a compression specimen whose length was three times the diameter, but it did not prove satisfactory. The shear specimen was 4 inches long and 0.750 inches in diameter and was subjected to a shearing force while being held in a suitable jig. No deformation instruments were attached to the jig or specimen. The scleroscope and Brinell hardness tests were taken on a polished cross-section of the specimen. On the softer specimens it was found that a more reliable value could be found for the Brinell reading by using a 1500 kilogram weight instead of either the 3000 kilogram or the 500 kilogram weight.

Foundry Practice

In order that a good piston be cast, it is necessary to use good raw materials. Since a piston is machined practically on all external surfaces and through the pin bosses, it must be free from blow holes and contain no hard spots due to segregation or other causes. Surface defects on the machined parts of a piston are easily detected and should be a cause for rejecting the piston. Since phosphorus must be low to avoid segregation, the pouring temperature should be somewhat higher for the thin-walled pistons than for ordinary castings.

The charge used for making the castings for the present test

consisted of the following: steel, 17½ per cent; charcoal pig, 55 per cent; scrap, 27½ per cent. The steel consisted entirely of railway splice bar of the following composition: carbon, 0.08-0.10 per cent; manganese, 0.35-0.60 per cent; sulphur, 0.06 per cent; phosphorus, not over 0.10 per cent. The charcoal pig had the following composition: total carbon, 3.50-4.25 per cent; manganese, less than 1.00 per cent; sulphur, 0.05 per cent; silicon, 0.75-1.50 per cent, phosphorus, 0.18-0.20 per cent. The scrap consisted entirely of foundry returns. The charge was melted in a cupola furnace with sufficient quantity of high grade coke and flux to give fluidity to the metal and produce a sound piston casting.

Heat Treatment

The pistons used in the machining study and the arbitration test bars of the E-series received exactly the same heat treatment, since they were heated simultaneously in the same furnace. For the arbitration test bars, Table I shows the temperature of heating in column A, the time required to attain that temperature in column B, the duration of time they were held at this temperature in column C and the duration of time in furnace while cooling in column D. In studying the machining properties of the pistons, no distinction was made for the small variation in heat treatment. The I-series of test bars was heat treated solely to study the effect on the mechanical properties, when the specimens were held at a given temperature for various lengths of time.

Table I

Specimens	A Heating Temp. Deg. Fahr.	B Time Coming to Heat Hours	C Time at Heat Hours	D Cooling Time Hours
E–13–1 and 2	1460	7½	½	10
E–14–5 and 6	1460	7½	½	10
E–16–1 and 2	1500	6	2	10
E–18–1 and 2	1500	6	2	10
I–25–0*	1650	2½	0	0
I–25–1*	1650	2½	1	0
I–25–2*	1650	2½	2	0
I–25–3*	1650	2½	3	0

*Two specimens.

All the other specimens used in this study were tested as received from the foundry. The critical temperature of this cast iron

was determined by the inverse rate method and found to be at 1430 degrees Fahr. All heat treated specimens were heated above this temperature and cooled slowly so that they received a treatment commonly known in steel practice as annealing.

Chemical Analysis

The chemical composition of some of the specimens show satisfactory agreement with the charge used as is evident from the results shown in Table II. The amount of combined carbon in the annealed specimens appears to be somewhat higher than is shown by observation of the photomicrographs. In the not-annealed specimens the agreement is more evident. The silicon, phosphorus and sulphur content are low and show that a good quality iron can be made into a thin casting.

Table II

Chemical Composition

Sample	T.C.	C.C.	Mn	Si	P	S	Heat Treatment
E-13-1..	3.65	0.32	0.65	1.22	0.29	.07	Annealed
E-13-3..	3.65	0.68	....	....	....	....	None
E-14-6..	3.46	0.24	0.69	1.31	0.34	.09	Annealed
E-14-7..	3.45	0.74	....	....	....	....	None
E-16-1..	3.13	0.15	....	....	....	....	Annealed
E-16-3..	3.16	0.65	0.59	1.66	0.30	.09	None
E-18-2..	3.58	0.16	....	....	....	....	Annealed
E-18-4..	3.56	0.84	0.57	1.22	0.22	.06	None
I-25-1..	3.52	0.30	0.53	1.72	0.22	.08	Annealed
I-25-9..	3.47	0.68	0.55	1.70	0.22	.07	None

Machining Properties

Cast iron pistons are usually machined from thin-walled castings having numerous ribs, webs, or partitions of various thicknesses. Not only are internal stresses set up in the casting because of the different rates of cooling of the various sections, but also segregation of the carbides, hard spots and brittleness are often found. These different conditions in the casting, together with the condition of the surface of the casting, nature of the cutting tool, and the kind of coolant used, govern the machinability of the specimen.

Depending on the time and temperature of the process, annealing will relieve the internal stresses and to a certain extent, elim-

inate the hard spots, break up the segregation and refine the structure. These all tend to improve the machining properties. In order to determine the advantages of the annealing process, experiments were conducted on the rough pistons that had been annealed with the arbitration bars, and the results similarly obtained on pistons that had not been annealed, are compared. The results are listed in Table III.

The determining factor of the test was a breakdown of the cutting edge of the tool. The breakdown was due to the striking or chopping effect of the out of round and uneven surface of the casting at the nose of the tool, causing the tool to break or push

Table III
Cutting Tests

	A = cutting speed in feet per minute	B = spindle speed	C = carriage speed in inches	D = back arm feed in inches	E = depth of cut in inches	F = width of cut in inches	G = average number of pistons machined in one setting, piston not annealed	H = same as G, except piston annealed	I = approximate ratio of the number of annealed pistons machined to those not annealed	J = tolerances in inches
Operation										
Rough Skirting	78	80	....	.027	....	.063	12	25	2	± .010
Rough Turning	156	160	.027	....	.027	.094	8	47	6	± .015
Finish Turning	156	160	.015	....	.015	.063	28	360	13	± .002
Roughing Ring Grooves	156	160	....	.027	....	.125	75	225	3	± .016 depth ± .016 width
Finish Ring Grooves	176	180	....	.005	....	.188	22	112	5	± .005 depth ± .0005 width

away from the work. All cuts were taken with a ground No. 3 stellite tool bit flooded with a heavy stream of specially prepared coolant.

The machining tests show a decided advantage in favor of the annealed castings. The ratios obtained in the rough skirting and rough turning of the annealed and unannealed castings are approximately 2 to 1 and 6 to 1, respectively. The average endur-

ance of the tool in the finish turning operation was in the ratio of 13 to 1 in favor of the annealed piston. This large variation was, no doubt, due to the favorable condition of the internal structure produced by the annealing process. In the rough turning of the ring grooves, the end point of the test gave the ratio of 3 to 1 in favor of the annealed casting. For the finish turning this ratio is 5 to 1.

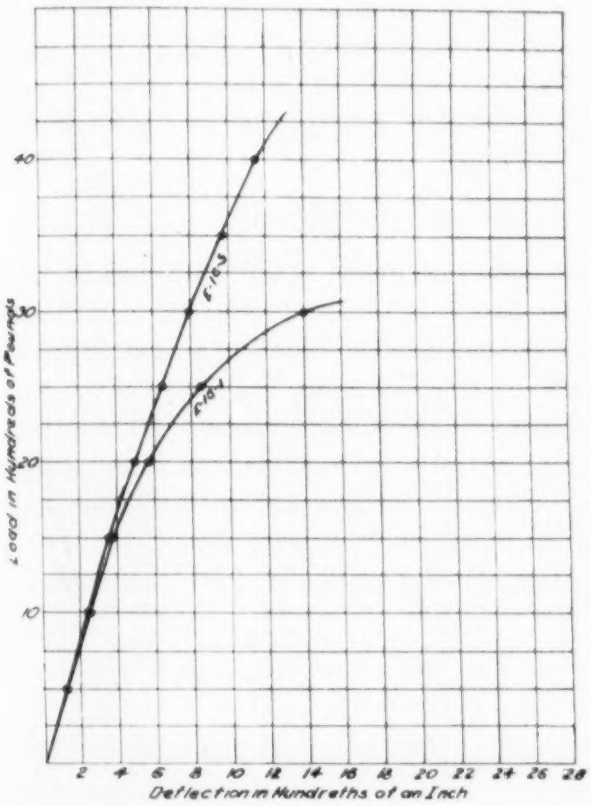


Fig. 1—Load-Deflection Curves from the Transverse Tests.

Transverse Tests

The results of the transverse tests are given in Table IV. The values of load and deflection (P and $\triangle$ respectively in the table) were determined directly from an autographic load-deflection curve. These curves are reproduced for specimens E-16-1 and E-16-3 in Fig. 1. The value of the modulus of elasticity, E, was calculated by drawing a line tangent to the load-deflection curve at the origin and using equation

$$E = \frac{P\ l^3}{48\ I\delta}$$

where P = an arbitrarily chosen load,
δ = the corresponding deflection for the load P as given by the tangent line,
l = the length of the span,
and I = the moment of inertia of the cross-section.

The value of the modulus of rupture, S_b, was calculated by the formula

$$S_b = \frac{Mc}{I}$$

where M = the maximum bending moment in the specimen at fracture,
and $\frac{I}{c}$ = the section modulus.

Since a cast iron specimen, when subjected to transverse load, breaks on the tension side of the specimen, it is desirable to compare the value obtained for the bending stress with that found for the tensile stress. The column headed S_b/S_t gives the ratio of the modulus of rupture to the tensile stress. This ratio varies from 2.19 to 2.84. Upton[1] gives a formula for calculating the relation existing between these two properties, but by this equation the ratio must have values between 1.00 and 2.00. Bach's[2] formula for this same relation gives a ratio of 2.05 for round bars. It is

$$S_b = K \sqrt{\frac{c}{y}} \; S_t$$

where K = a constant depending on the cross-section form,
c = distance from the neutral axis to the outer fiber,
and y = distance from the neutral axis to the centroid of the area under tension.

His experimental results upon tension and bending bars which have been machined gives values somewhat higher than 2.05. It is to be noted that the transverse bars used in the present study were not machined and the influence of the casting skin on the properties has not been determined.

The heat treatment of the test specimens has decreased the bending strength about 29 per cent and the modulus of elasticity about 9 per cent. While the average values of the deflection show

[1]Upton, Materials of Construction, page 79.

[2]Bach, Elastizität und Festigkeit, page 293.

a decrease, it is comparatively small. The variation of the individual values is greater than this decrease. The heat treated bars show a greater increase in deflection with the same load increment than do the untreated bars, especially after being loaded above 1500 pounds. It was noticed in performing the tests that the report or sound of the fracture of the heat treated bars was barely perceptible or dull while for the others it was very distinct and snappy. This would indicate that the material had lost much of its brittleness in heat treatment. This same phenomenon was observed by Potter[3]. It is to be noted that all transverse specimens satisfy the specifications as required by the American Society for Testing Materials.

Table IV

No.	E_b Modulus	P Load	$\triangle$ Deflection	S_b Modulus of Rupture	S_b/S_t Ratio
E-13-1	11,550,000	2700	.140	42300	2.80
E-13-2	10,900,000	2750	.130	43000	2.84
E-13-3	13,350,000	3900	.130	61000	2.61
E-13-4	12,250,000	3300	.120	51700	2.13
E-14-5	13,050,000	2550	.100	39900	2.42
E-14-6	12,250,000	2850	.125	44600	2.78
E-14-7	13,950,000	4200	.130	65600	2.81
E-14-8	13,050,000	3700	.130	57800	2.53
E-16-1	12,000,000	2900	.130	45300	2.81
E-16-2	12,250,000	2750	.105	43000	2.62
E-16-3	13,050,000	4350	.130	68000	2.21
E-16-4	13,650,000	3800	.110	59400	2.75
E-18-1	13,050,000	2900	.110	45300	2.81
E-18-2	12,500,000	2650	.125	41400	2.52
E-18-3	14,300,000	4000	.135	62600	2.36
E-18-4	13,500,000	3700	.100	57800	2.19
Ave. Heat Treated ...	12,200,000	2760	.120	43100	2.70
Ave. Not Heat Treated ...	13,380,000	3870	.124	60500	2.45
Ratio	.912	.713	.968	.713	....

Tensile and Hardness Tests

Table V shows the results of tensile and hardness tests and Fig. 2 shows the tensile stress-strain curve for specimens E-16-1 and E-16-3. The curves for the other heat treated specimens were

[3]Potter, *Transactions*, American Foundryman's Association, 1925.

practically identical to the one for E-16-1, and those not heat treated about the same as for E-16-3. The stress-strain diagram shows practically linear proportionality between the stress and strain for at least a load up to 5000 pounds per square inch and from this relationship there has been computed the modulus of elasticity as given under column E_t of this table. The modulus of elasticity of the heat treated specimens has been decreased about

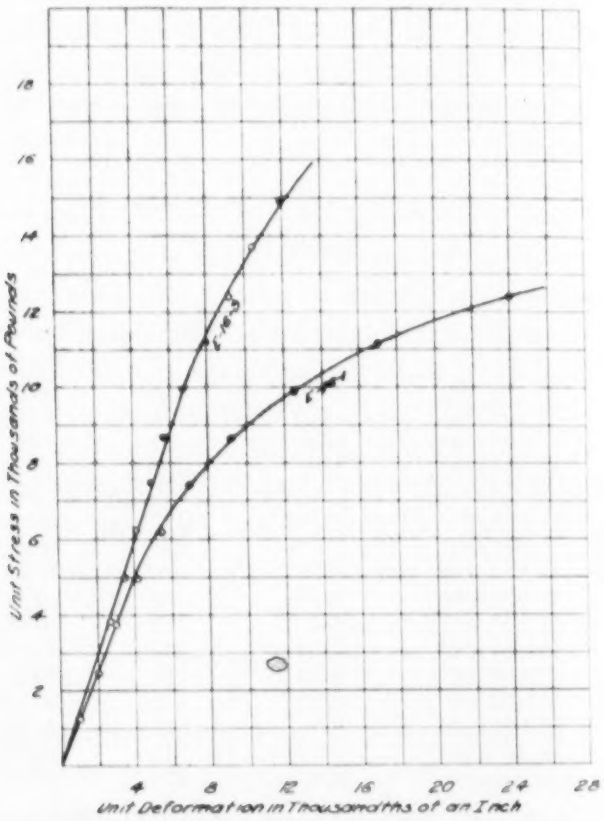


Fig. 2—Stress-Strain Curves from the Tension Tests.

8 per cent in comparison to the value of those not heat treated. The average value of this property in tension is a little higher than that obtained from the transverse test as is to be expected.

The tensile strength of the heat treated specimens is decreased by about 36 per cent as compared to the values of the not heat treated specimen. For the Brinell and scleroscope hardness this same decrease is respectively 38 per cent and 41 per cent. This indicates that the hardness test for this type of cast iron furnishes a good comparison of the relative tensile strength of the heat treated to not heat treated cast iron.

Table V

	E_t	S_t	Brinell	Shore
E-13-1	12,500,000	15100	109	21
E-13-2	12,500,000	15100	104	20
E-13-3	12,500,000	23300	179	34
E-13-4	12,900,000	24300	170	35
E-14-5	12,500,000	16500	114	21
E-14-6	12,100,000	16700	109	22
E-14-7	13,200,000	23600	170	32
E-14-8	12,600,000	22800	174	34
E-16-1	12,500,000	16100	104	19
E-16-2	12,500,000	16400	104	20
E-16-3	14,700,000	30800	179	35
E-16-4	14,700,000	21600	179	36
E-18-1	12,500,000	16100	109	21
E-18-2	12,500,000	16400	106	19
E-18-3	14,700,000	26500	170	33
E-18-4	13,200,000	26300	170	33
Ave. Heat Treated	12,500,000	16050	107	20
Ave. Not Heat Treated	13,560,000	24900	174	34
Ratio	.922	.644	.616	.588

Compressive and Shear Tests

The results obtained from compression and shear tests are given in Table VI. In Fig. 3 is shown the stress-strain relation for specimen E-16-1 and E-16-3 when tested in compression. Here again the values of the stress and deformation show a straight line relation up to a stress of about 10,000 pounds per square inch for the heat treated specimens and somewhat higher for those not heat treated. Also the relative deformation shows a more rapid increase for the former than for the latter specimens after passing this straight line relation. From this proportional relationship the modulus of elasticity in compression was determined. This value is tabulated in Column E_c. From the compressive tests the modulus of elasticity of the specimens showed a decrease of about 9 per cent due to heat treatment.

The compressive stress is tabulated in column S_c and the relative decrease of this property for the heat treated specimens is about 40 per cent. After the specimen failed the angle of the plane of failure was measured and tabulated in column Φ. The shearing stress is given in column S_s and the decrease here is about 36 per cent. In the column Θ there has been tabulated the

value of the angle of failure for cast iron as computed by the formula[4].

$$\cot \Theta = \frac{2\ S_s}{S_c}$$

where S_s = the maximum unit shearing stress

S_c = the maximum compressive stress.

It is to be noted that the computed and measured angles of failure are in good agreement. It is, therefore, possible to obtain a satisfactory value for the shearing stress if the compressive stress and the angle of failure have been determined.

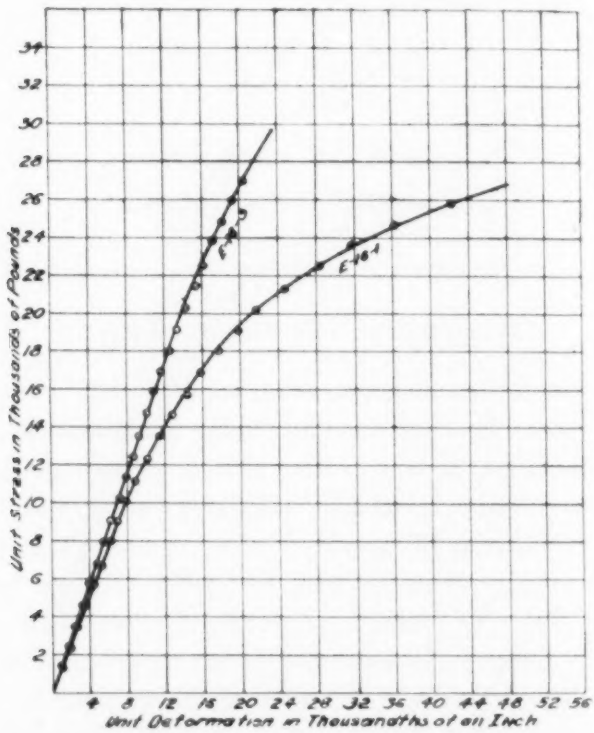


Fig. 3—Stress-Strain Curves from the Compression Tests.

Microstructure

A study was made of the microstructure of all the test bars whose mechanical properties were determined. Low and high power magnification were used to bring out the true structure of the specimens.

Photomicrographs, Figs. 4, 5 and 6, show the typical structure

[4]Upton, Materials of Construction, page 84.

found in the not-annealed specimens. Photomicrograph Fig. 4, taken at 100 diameters, shows clearly the even distribution of the graphite particles throughout the ground mass which is unresolved at this magnification. A gray cast iron of this structure shows sat-

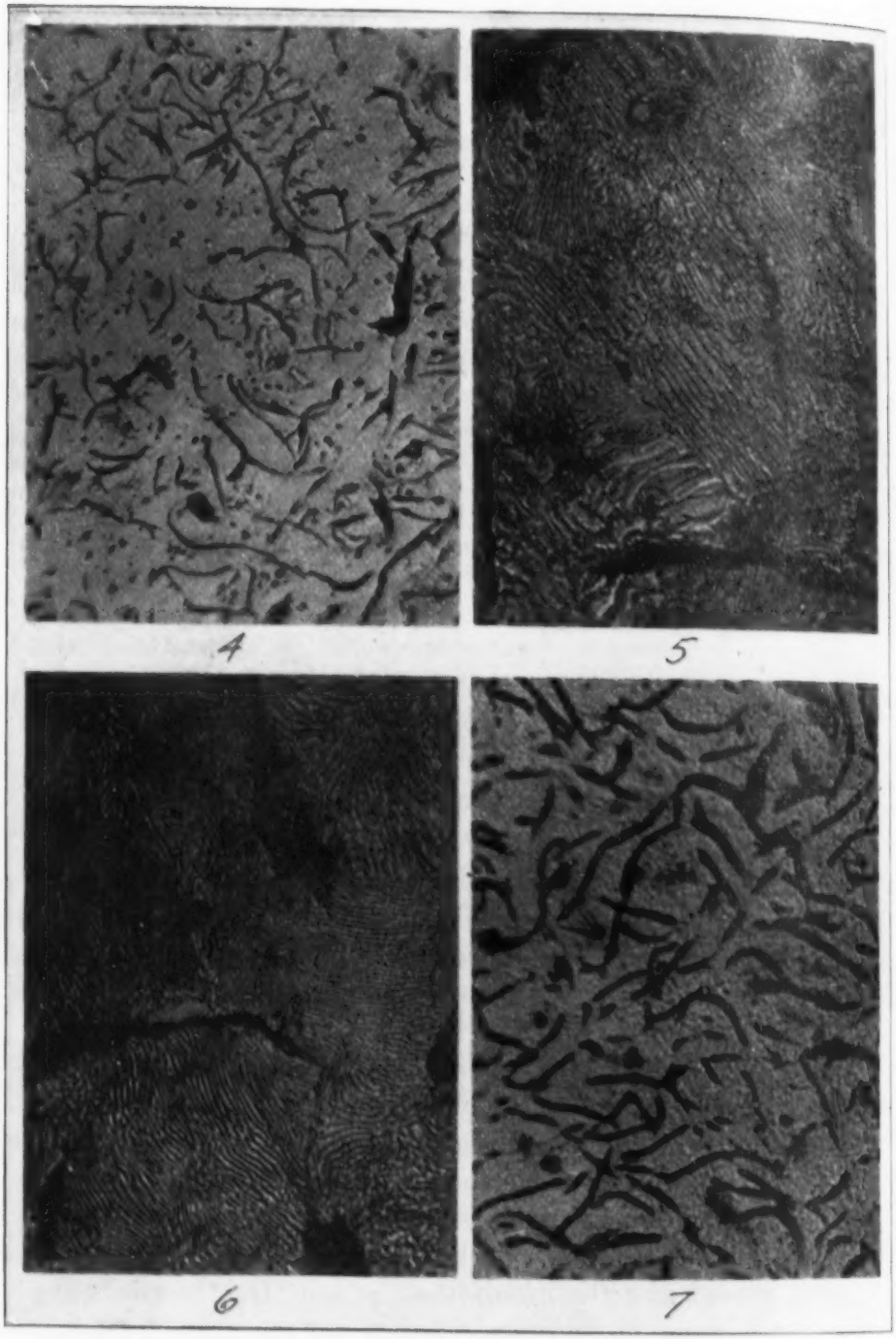
Table VI

	E_c	S_c	Φ	S_s	θ
E–13–1	12,100,000	60600	56°	19700	57°
E–13–2	13,500,000	59000	57°	19800	56°
E–13–3	14,350,000	97500	58°	28900	60°
E–13–4	14,200,000	100700	58°	29500	59°30′
E–14–5	13,950,000	63200	56°	21400	56°
E–14–6	12,750,000	62600	57°	22100	55°
E–14–7	13,900,000	98500	58°	32300	56°30′
E–14–8	12,050,000	97800	58°	31600	57°
E–16–1	13,800,000	57500	56°	21200	54°40′
E–16–2	12,700,000	60800	57°	20900	55°30′
E–16–3	15,000,000	113000	56°	38600	55°30′
E–16–4	15,050,000	103000	58°	32100	58°
E–18–1	12,250,000	60600	57°	20900	55°30′
E–18–2	11,500,000	60300	57°	21400	54°30′
E–18–3	13,800,000	100700	57°	33400	56°30′
E–18–4	14,000,000	101200	57°	33600	56°30′
Ave. Heat Treated	12,820,000	60600	56°30′	20900	55°30′
Ave. Not Heat Treated	14,040,000	101500	57°30′	32500	57°30′
Ratio	.913	.597		.643	

isfactory casting practice. The magnification at 1000 diameters is reproduced in photomicrographs, Figs. 5 and 6. With this magnification, there were found small areas of cementite near the graphite flakes. The ground mass was a fine-grained pearlite having a curly like structure.

Annealing this cast iron above the critical temperature produced a complete change in the microstructure. Photomicrograph, Fig. 7, taken at 100 diameters, shows a decided increase in the size of the graphite particles. The increased magnification (1000 diameters) in photomicrographs Figs. 8 and 9 shows more effectively the increased size of the flakes. The fine-grained pearlitic structure is entirely absent and the groundmass is now free ferrite. Deep etching of the annealed specimens brings out the grain boundaries of the ferrite when subjected to high magnification.

In the annealed specimens, the combined carbon is nearly all eliminated. This elimination of the combined carbon reduces the hard spots in the casting and gives it excellent machining properties.



Figs. 4, 5 and 6—Photomicrographs Showing the Typical Structure Found in an Unannealed Specimen. Fig. 4—Specimen E-16-3 was etched in Picric Acid. Mag. 100X. Fig. 5—Specimen E-16-3 was Etched in Picric Acid. Mag. 1000X. Fig. 6—Specimen E-18-3 was Etched in Picric Acid. Mag. 1000X. Fig. 7—Shows the Structure of a Cast Iron Similar to Figs. 4, 5 and 6 After Annealing Above the Critical Temperature. A Decided Increase in the Graphite Particles Will be Observed. Mag. 100X.

Effect of Time on Annealing

In Table VII are given the results which show the effect of holding the specimen to be heat treated at a temperature of 1650 degrees Fahr. for 0, 1, 2 and 3 hours, specimen 1-25-9 was tested as received from the foundry. All others were heat treated. About the same charge was used in making castings as in the case of the E-series.

It is of interest to note that the specimen 1-25-9 had about the same mechanical properties as the average values of the not heat

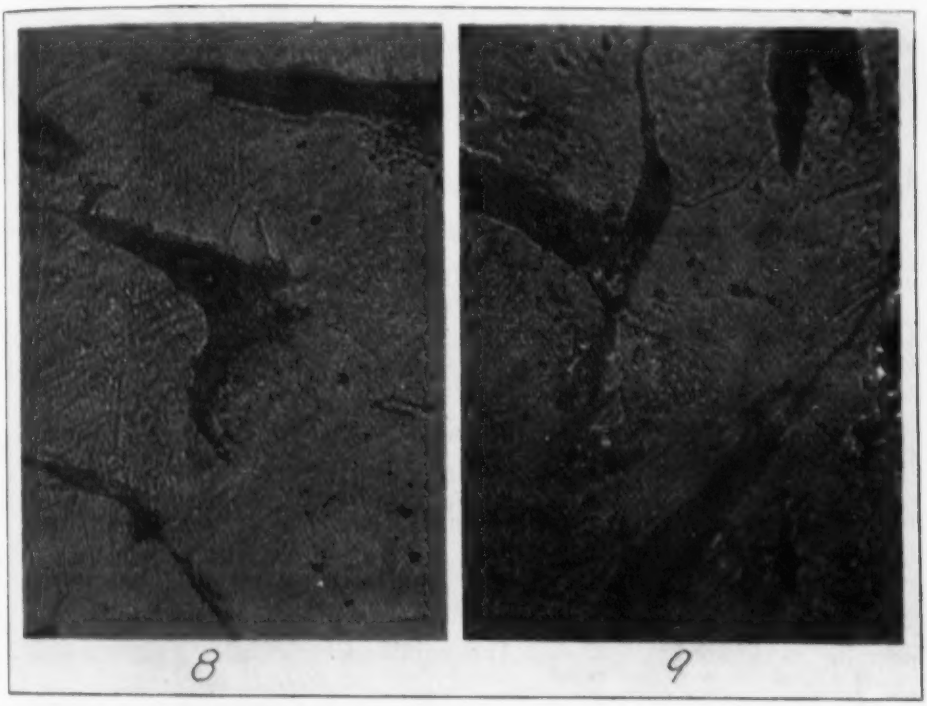


Fig. 8—Photomicrograph of Specimen E-16-1 After Annealing Above the Critical Temperature. Specimen Etched in Picric Acid. Mag. 1000X.

Fig. 9—Photomicrograph of Specimen E-18-2 After Annealing Above the Critical Temperature. Specimen Etched in Picric Acid. Mag. 1000X. The Marked Increase in Size of the Graphite Flakes Will be Noted.

treated specimens of the E-series, except in the case of the modulus of elasticity. The value for this specimen is a trifle higher.

The length of time the heat treated specimens were held at 1650 degrees Fahr. appears to decrease the modulus of rupture and the total deflection. The other properties do not seem to be affected to any appreciable degree by holding them at this temperature for an interval of time up to 3 hours.

The modulus of rupture for the heat treated specimens E-25-0

show an appreciable decrease in comparison to the not heat treated specimen and continues to decrease as the interval of heating increases. In the case of the deflection, these same specimens show an increase in deflection as compared above and then continue to decrease in the same manner as for the modulus of rupture.

Table VII

	Bending				Tension		
	P_b	$\triangle$	S_b	E_b	S_t	S_b/S_t	E_t
I–25–9 ..	3800	.120	59,400	14,500,000	24,500	2.42	14,600,000
I–25–0* ..	3100	.137	48,400	13,600,000	17,100	2.78	12,700,000
I–25–1* ..	3025	.129	47,200	13,800,000	16,700	2.82	13,250,000
I–25–2* ..	2825	.110	44,900	13,250,000	16,300	2.75	13,500,000
I–25–3* ..	2550	.085	40,000	13,150,000	16,300	2.45	13,200,000

	Compression			Hardness	
	S_c	Φ	E_c	Brinell	Shore
I–25–9	100,500	58°	14,800,000	179	35
I–25–0*	65,700	58°30′	13,500,000	112	22
I–25–1*	63,600	57°30′	13,100,000	109	21
I–25–2*	64,700	56°	13,000,000	106	21
I–25–3*	64,900	58°	12,800,000	106	22

*Average of two specimens

Hardness of Automobile Pistons

Automobile pistons used in commercial vehicles vary widely in Brinell hardness as is shown in Table VIII. In the case of used pistons, hardness values were found to vary from 187 to 103. As new pistons were not available in all cases it is impossible to

Table VIII

Automobile Co.	Piston No.	Used	Automobile Co.	Piston No.	New	Used
1	1	187	11	1	182	
2	1	179	11	2		172
1	2	178	11	3		138
6	1	170	11	4		122
1	3	158	11	5		113
6	2	151	12	1	120	
10	1	148	12	2	108	
4	1	142	12	3		112
9	1	140	12	4		108
5	1	135	13	1	182	
5	2	128	14	1	114	
7	1	108				
8	1	108				
9	1	103				

decide what caused this variation in hardness. In the case of automobile company 11, a new piston had a hardness number of

182, while for the used pistons the hardness number varied from 172 to 113. If the used pistons were originally as hard as the corresponding new piston (there is every reason to believe they were), then the decrease in hardness in the used pistons was due to the temperature of the piston while operating in the engine under commercial conditions. Huebotter[5] states that the temperature at the center of the head of a gas engine piston under full load is about 850 degrees Fahr. Under extraordinary conditions this temperature may be exceeded. And it is possible that the temperature could be sufficiently high to produce the decreased hardness. In the case of some of the used pistons, their hardness was equal to the value of new pistons; and since no data were available on the time or condition of service of these pistons, it would be difficult to decide whether or not the operating temperature effected their hardness.

Summary

The results recorded in the previous pages may be summarized as follows:

A. In heating a gray cast iron above the critical temperature and cooling slowly:
   (1) The machinability is benefited in the ratio of approximately 6 to 1.
   (2) The maximum strength properties and hardness is decreased from 30 to 40 per cent.
   (3) The modulus of elasticity is decreased from 8 to 10 per cent.
   (4) The maximum deflection under transverse loads is only slightly decreased.
   (5) The allowable working stresses of cast iron are decreased about 10 per cent for the same deformation.
   (6) The microstructure is materially changed.

B. The commercially used pistons vary in hardness from those un-annealed to those annealed above the critical temperature.

Acknowledgments

This investigation owes its inception to F. C. Bahr of the Arrow Head Steel Products Company of Minneapolis. The above

[5]Huebotter, Mechanics of the Gasoline Engine.

company furnished the specimens, heat treated, and machined the material; and through the courtesy of their Messrs. Wilbert, Thill, and Emmons, it was possible to carry on the tests. Acknowledgment is due Miss Green of the metallographical laboratory of the University of Minnesota for her assistance in preparing the photomicrographs. The mechanical tests were performed in the experimental engineering laboratories of the University of Minnesota.

The Authors

George C. Priester was born February 1, 1885, near Gutenberg, Iowa. He graduated with the degree of Bachelor of Engineering from the State University of Iowa in 1910 and received the degree of Master of Science from the University of Minnesota in 1916. Since

G. C. PRIESTER

F. J. CURRAN

1910 he has been teaching at the University of Minnesota in the department of mathematics and mechanics and is at present an associate professor in this department.

Mr. Priester is a member of the American Society for Testing Materials, American Society for Steel Treating, Sigma Xi, Tau Beta Pi, Minneapolis Engineers' Club, Society for the Promotion of Engineering Education and Theta Xi.

He is a co-author of a paper "Effect of Temperature on the Mechanical Properties of Steel", published

in Chemical and Metallurgical Engineering January 17, 1923, and author of "Determination of Poisson's Ratio", Minnesota Technolog, May, 1925.

FRANCIS J. CURRAN was born August 6, 1902, in Duluth, Minnesota. He graduated with the degree of Metallurgical Engineer in 1924 from the University of Minnesota. He was engaged as metallurgical engineer for two years with the Arrow Head Steel Products Company of Minneapolis and at the present time is metallurgist for the Associated Chemists, Minneapolis, Minnesota.

He is a member of the American Society for Steel Treating and Sigma Rho.

DISCUSSION

Written Discussion: By H. A. Schwartz, manager of research, National Malleable and Steel Castings Co., Cleveland.

The authors have shown very plainly the changes of physical properties resulting in a gray iron of approximately eutectoid matrix, which is graphitized substantially to completion. It may be well in this connection to call attention to the fact that the metallographic change which they produced did not take place altogether at 1650 degrees Fahr., but in part occurred during an appropriately slow cooling, below the critical point. It is only below A_1 stable that ferrite alone can be present, under equilibrium conditions, in contact with carbon or iron carbide.

It has long been known that the malleable casting is the most machineable commercial ferrous product. The heat treatment of gray iron, as outlined by the present authors, is that normal in the malleable industry, but suitably abbreviated as required, for the purpose of completing the graphitization, which had largely progressed during the freezing of the original alloy. Under these conditions, the authors should have, and actually observed, a corresponding improvement, with regard to the machineability.

Their product differs from that of the malleable foundry in that such a heat treatment as theirs, while it can accomplish complete graphitization, cannot produce the particularly desirable nodular form of carbon secured when the original material is completely free from graphite.

Thus, for a Brinell number on their annealed samples, slightly lower than that of ordinary malleable, they observe an average tensile strength of 16,000 pounds per square inch, instead of 50,000 pounds or over, normal to malleable iron. The lower deflection in cross bending of annealed gray iron specimens as compared with what might be expected from malleable is similarly shown.

The commercial utility of the heat treatment of gray iron would seem to depend largely upon whether machineability or strength is the property most valued in the resulting product. As the authors have shown, the heat treatment of gray iron improves the former at the expense of the latter. The

engineer or heat treater confronted with this problem should, therefore, consider carefully, whether the casting made more machineable by the correct heat treatment will still be useful for the intended purpose. In many cases, this will possibly not be true. For example, we may consider the cylinder block or the piston ring of the automotive engine, where resistance to wear in use should be of much greater value than the decreased cost of production accompanying greater machineability.

The preceding comments are not made by way of criticism or amendment of the authors' paper, but rather in the hope that they may be of some slight interest to readers of that paper, in making applications in their own engineering practice.

R. S. MACPHERRAN: I merely wish to ask a question as to the record, if any, kept on the relation between the wear of the unannealed and the annealed piston. In our own experience we are asked to keep the iron hard enough to show satisfactory wear. The authors' results of 20 scleroscope or 105 Brinell would seem to us to be soft for a casting to be subjected to wear.

Another fact I noted was the lowness of the tensile value. This iron runs 1¼ to 1½ per cent in silicon with 17 per cent steel, 5 per cent charcoal pig iron, yet the tensile strength only goes up to less than 25,000 pounds per square inch. We expect to get well over 30,000 pounds per square inch with the same analysis iron. We made quite a number of experiments on annealed castings, but never went to the extent the authors did. We never like to lose over 15 to 20 per cent of the tensile values.

W. J. MERTEN: It seems to me that this paper should stimulate some additional work on annealed iron particularly with reference to the accuracy of the testing for the properties which the authors have described. It does not seem justified to conclude that the actual values do change in the different samples. This is particularly true with reference to the moduli of elasticity, and I would like to ask what make of testing machine and method was employed, to determine it. The variations are so slight that I am quite convinced they are errors in testing rather than material differences.

I do not believe the modulus is changed by the annealing operation, and the fact that it is higher one time and lower the next is certainly not convincing as to its accuracy. Additional tests to verify these values should be made especially in view of the fact that we have no definite or accurate figures on the properties of cast iron.

DR. S. L. HOYT: I notice the Brinell hardness is given as low as 108. It occurs to me this is quite unusual for cast iron. I would like to be informed if this is apt to be met with in pistons, either new or old, or was this the exception?

DR. D. J. MCADAM, JR.: The annealing has been carried so far that the microstructure contains no pearlite. The general properties of cast iron are better for most purposes if the matrix contains some pearlite.

CHAIRMAN T. D. LYNCH: One day last week I saw two samples of high-test iron 10 inches long by approximately ⅝-inch in diameter that had been normalized. The normalizing operation caused an elongation of 1/16-inch on an inch.

I am wondering if the softening tendency of these treatments are not due to the expansion of the metal and the corresponding opening up of texture. In the case of rings in service, they expand and wear off. I noticed in some of the rings of long time in service that the hardness became less as the age increased. This suggestion is offered in the hope someone may wish to discuss it further.

J. STURROCK: In connection with the wearing qualities of cast iron castings, Ingersoll-Rand Company at the Athens plant have carried on a considerable number of experiments along this line. We have had considerable trouble with hard castings, and, in trying to anneal them, find that there is not much change in hardness until the temperature reaches the critical temperature and then the structural change is so quick that it is impossible to stop at a point which will give a desirable 0.70 per cent combined carbon. The usual result is complete graphitization of the carbon, resulting in nothing but carbon and ferrite which gives a very poor wearing surface. On account of these results, it was necessary for us to discontinue annealing cast iron castings.

G. C. PRIESTER: I will answer some of the questions as far as I am able. We did not conduct any wearing tests on it from a laboratory standpoint, so that the comparison between the annealed and the unannealed, as would be indicated by the wearing of the piston, is not known.

I might add that I just came in contact a short time ago with a company which is using an anneal up to 1650 degrees Fahr. on their bearings, cast iron bearings. They are using cast iron bearings in preference to any other bearings they have been able to use in their work. This is high speed machinery. Whether that has any significance I do not know. They were continually getting failures in all kinds of bearings, and finally they found cast iron which was annealed up to 1650 degrees Fahr. was the most satisfactory bearing.

I also want to say that the low tensile strength was probably due to the kind of casting that was to be poured.

Making a piston casting is the foundryman's job and he has got a big job on his hands. The casting must be machined on all sides, and it is subjected to inspection after it is machined on all the surfaces. It is also a thin casting. You will note in the chemical composition a low phosphorus and low sulphur content. He is pouring a thin casting. Those are problems that the foundryman must meet. You can probably do better work, get better strength if you have a different casting to make. I merely say that, not from my standpoint, but from the foundryman's standpoint.

Also, in regard to the testing apparatus, I want to point out that there were four sets of specimens, 13, 14, 16 and 18. These were all taken from different heats on different days, so that you would expect a variation. Also, the specimens were taken from a production foundry and not from a laboratory foundry where you can make identically the same thing. They were probably making different size pistons on one day than they made on the day the next specimens were taken, and so on. So that you would expect a variation in them. In all my testing of cast iron I have found cast iron to vary from the values that I get, say, 10,000,000 up to 20,000,000, and I have seen results where the

values run up as high as 30,000,000 for the modulus of elasticity, determined in the way we determine it, so you can see there is wide variation there, depending on the foundry properties.

Now, the low Brinell hardness, in answer to Dr. Hoyt's question, I might say that in making our tests on the low Brinell hardness, we used 1500-kilogram weights and not 2000. We did not use the 500-kilogram weights because we found that the indentations were unreliable and hard to measure. On the thin castings we had to support the piston in such a way that it would not crack under the load, and we found that a 1500-kilogram load with proper support would give very good compression to make a good reading and also a check reading or check readings. That is, we did not depend on a single reading.

I am not prepared to say much about or answer Dr. McAdam's question, except to say that the annealing did give us this structure. Whether that is the desirable structure for automobile pistons, I am not prepared to say. Our object was to run tests on physical properties of annealed cast iron and this company was willing to put out the pistons under those conditions. What we might suggest as a study, is the effect of the properties when the temperature is not carried as high as we carried it. I think it would bear fruitful results.

STEEL FURNACE CRYSTALS

By J. S. G. Primrose

Abstract

From a description of a steel-furnace-bank crystal group, and micrographs of a section taken from a fifty-pound specimen, the striking features arising from the slow cooling of hyper-eutectoid steel are discussed with reference to the brittle Widmanstätten structure. The very low power resolution of the pearlite lamellae; and the intersection of the primary cementite needles are noteworthy.

WHEN small pockets of molten steel are entrapped on the banks of open-hearth furnaces and, instead of working down to the usual low carbon point of the bath, are left with moderately high carbon contents to cool down slowly in position in the furnace, some interesting results in the formation of massive steel crystals often occur. When they come to be broken up the appearance of the fractured faces is often peculiar, and quite extensive conchoidal surfaces are formed in this way with quite smooth facets many inches across, although

Fig.1—Portion of Mass of Steel Crystals Slowly Cooled on Furnace Bank.

A paper by J. S. G. Primrose, metallurgical engineer, Manchester, England.

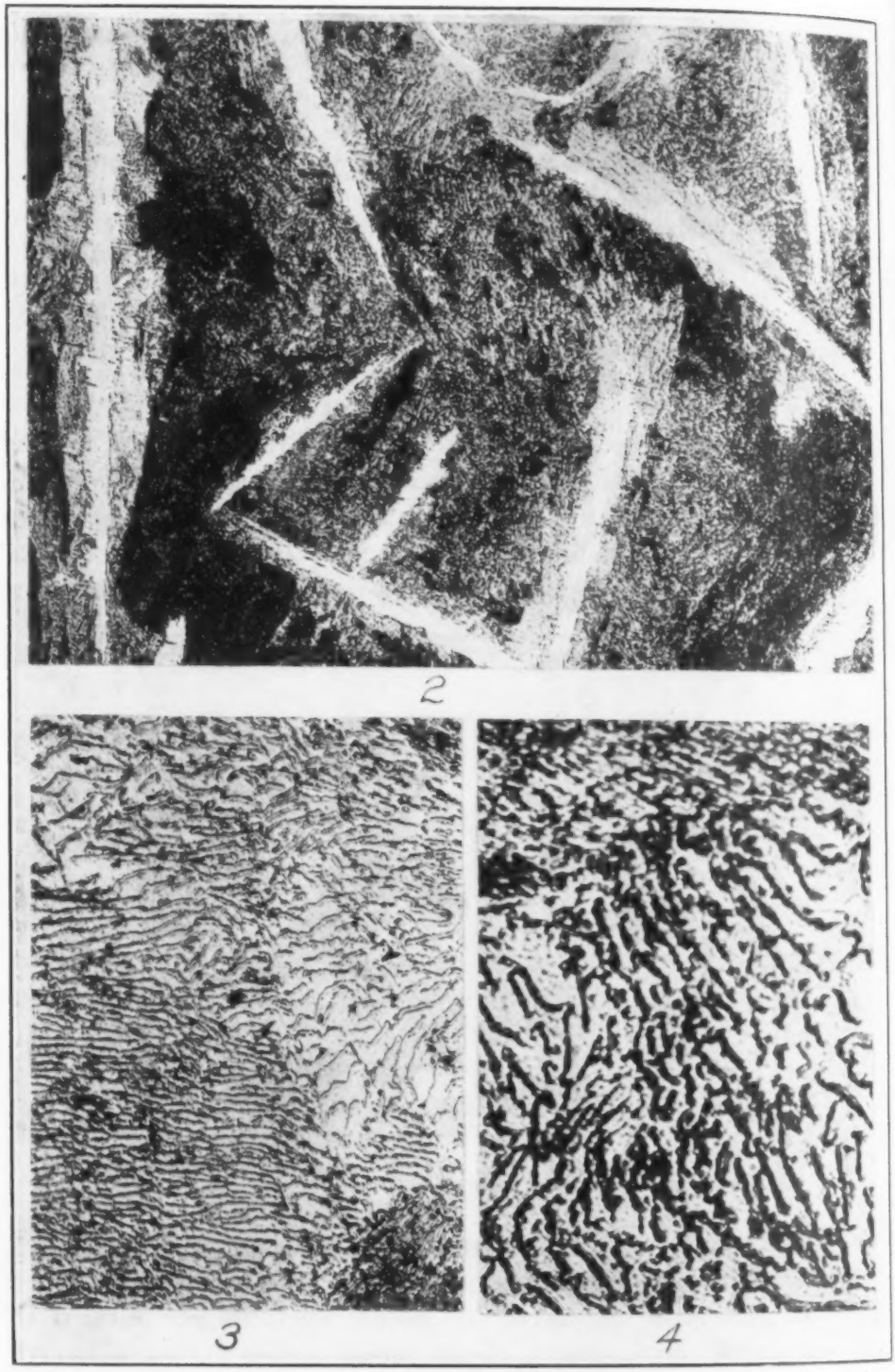


Fig. 2—Cementite Needles Lying in Four Directions in Pearlite Groundmass. Magnification 50 x. Fig. 3—Variously Oriented Grains of Pearlite in Steel Crystal. Magnification 200 x. Fig. 4—Two Differently Inclined Grains of Pearlite in a Steel Crystal Magnified 600 Diameters.

what might be termed the roots of the steel mass next to the lining may be comparatively small-faced and variously oriented.

A small portion weighing seven pounds, broken from a much larger piece approximately fifty pounds in weight, is shown considerably reduced in Fig. 1, and the features noted above, which must have been observed by many open-hearth furnace workers,

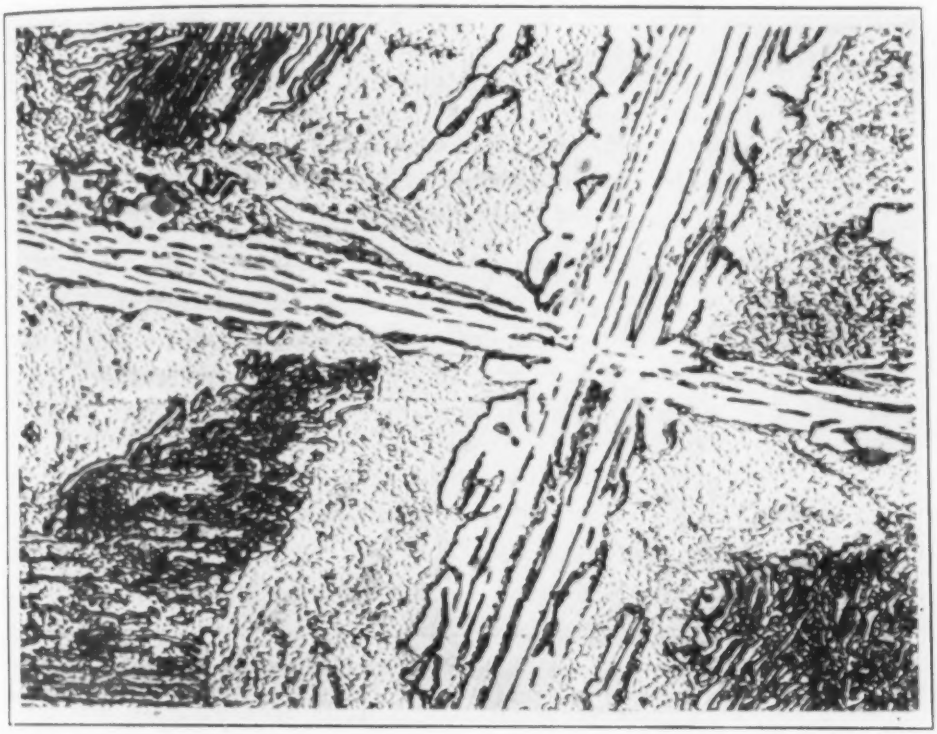


Fig. 5—Two Crossing Needles of Cementite in a Steel Crystal. Magnification 600 x.

are clearly visible. What is not so generally considered however, is the nature of the inner structure of this specially cooled piece of metal, and its study under the microscope often presents striking features which are worthy of consideration.

At a comparatively low magnification, the structure of the polished section reveals a remarkable series of variously oriented spikes of cementite, as in the case of a 1.30 per cent carbon steel crystal recently examined by the writer, and these needles quite obviously follow four different directions as befits the general section of the Widmanstätten structure. The pearlite grains also lie at highly variant angles to the octahedral plane, and thus cut the plane of the micro-section in some cases at high angles, roughly

80 degrees, and in this way give rise to laminations of ferrite and cementite which are situated so far apart on the surface as polished that they are perfectly distinguishable at such a moderate magnification as 200 diameters. The part of this specimen shown in Fig. 4 at 600 diameters is a small part near the same spot as the top portion of Fig. 3 shown at 200 diameters and from this it appears that one of the pearlite grains has a much greater inclination to the section than the others, which, from the width of the lamellae, appear to be much more nearly normal to the polished face.

Perhaps the most remarkable point in connection with the examination of this particular crystal was the finding of a part of the pearlitic area in which two distinct cementite needles actually cross, and appear to have interpenetrated one another in their growth. This is shown in Fig. 5 at 600 diameters magnification where the two needles cross each other at an angle slightly less than 90 degrees, the four corners made by their intersection being filled in with laminated pearlite.

NOTES ON THE A_3 STABLE TRANSFORMATION

A Contribution to the Subject of Graphitization

By H. A. Schwartz

Abstract

The present paper is not submitted as a conclusive report on a finished piece of work. It seems desirable to record certain, rather surprisingly consistent observations, even though they lead to a distinctly unusual conclusion. The observed facts and the conclusion, that A_3 stable is distinctly below A_3 metastable, are in the author's judgment adequately supported by evidence.

The suggestion that A_3 stable approaches A_2 in pure iron-carbon alloys, as the concentration of the element promoting graphitization decreases, is put forward as speculative only, although it seems to follow logically from the experimental work at our disposal.

Such a conclusion would have a bearing on the characteristics of solid solutions, and hence, would appear worthy of thought, and discussion by others.

Introduction

THE writer has previously shown in co-operation with others[1] that the alpha-gamma transformation takes place at lower temperatures in the stable system, iron, carbon, silicon, than in the metastable. Hayes[2] and his associates have held that this is not true, postulating that silicon is without effect upon the A_3 point. This supposition is not considered valid, and the most recent work of Wever[3] may be cited in rebuttal.

The purpose of the present paper is to present certain experimental data from which, by extrapolation we may draw conclusions as to the locus of the A_3 line in the stable equilibrium diagram.

Experimental

There being no known hypoeutectoid stable iron-carbon alloys, the location of A_3 by direct experiment is impossible. An

[1] "Effect of Silicon on Equilibrium Diagram of System Carbon-Iron Near Eutectoid Points," H. A. Schwartz, H. R. Payne, A. F. Gorton, *Transactions* of American Institute of Mining and Metallurgical Engineers, Vol. LXIX, 1923, page 791.

[2] "The Carbon Content of Pearlite in Iron-Carbon Alloys Containing One Per Cent Silicon," Anson Hayes and H. U. Wakefield, Transactions, American Society for Steel Treating, August, 1926.

[3] "Zur Allotropie des Eisens," Franz Wever, *Stahl und Eisen*, Vol. 45, Part 28, p. 1208, 1925.

A paper presented before the eighth annual convention of the Society, Chicago, September 20 to 24, 1926. The author, H. A. Schwartz, member A. S. S. T., is manager of research, National Malleable and Steel Castings Co., Cleveland.

indirect method of attack must, therefore, be adopted. The effect of silicon on the locus of the eutectoid point, both in concentration and temperature, has been experimentally determined under the writer's supervision, loc. cit.

The effect of silicon on A_1 may be regarded as the sum of its effect on the locus of the A_3 and on the carbon concentration of the eutectoid, and A_3 obviously ceases to have any separate existence apart from A_1 at concentrations of silicon above that at which the carbon concentration of the stable eutectoid becomes zero. If then, we observe the effect of silicon on A_1, at still higher concentrations, i. e., at values where there is no longer a change of carbon content of the eutectoid possible, the effect of silicon on A_1 and A_3 is identical.

Such an observation showed the stable A_1 point at 8.0 per cent silicon to be 1724 degrees Fahr. (940 degrees Cent.), as compared with 1589 degrees Fahr. (865 degrees Cent.) at 3.75 per cent silicon, by extrapolation in our former paper. This being at the rate of 17.6 degrees Cent. per 1 per cent silicon, as compared with 16.6 degrees Cent. per 1 per cent silicon, as previously observed in metastable alloys for silicon values above 1.2; i. e., in the range where A_2 is below A_1.

It was shown by the author in discussion of Hayes' and Wakefield's paper, loc. cit., that above this concentration the effect of silicon upon the carbon concentration of pearlite is small. By extrapolation from the data just quoted, A_3 stable approaches the value 1470 degrees Fahr. (799 degrees Cent.) as a limit when the carbon and silicon content approach zero. If the same system of experiment and reasoning were applied to other ternary systems of carbon and iron, the same limiting value of A_3 should be obtained, since the concentration of the third element is zero in the limit.

Walter states that boron in small amounts promotes graphitization. It cannot be employed, for our present purpose, for in larger amounts it inhibits graphitization[4], and hence, we cannot extend its investigation into regions where it entirely excludes carbon from the stable solid solution. Nickel may be used as a substitute for silicon, and also four other elements which the writer is unfortunately not permitted to name at this time. Among these

[4] "Chemical Elements Inhibiting Graphitization," H. A. Schwartz and G. M. Guiler, 1925 Convention of American Foundrymen's Association.

five, some raise and some lower the critical points, while one has only a slight effect. Two have little, or a negative effect in decreasing carbon solubility, while three have a positive effect, but less than that of silicon. All five elements have been studied in some detail, though not, of course, as thoroughly as the common element, silicon.

The values of Ac_3 stable, obtained by the same methods of extrapolation employed above, were—

	Degrees Cent.	Degrees Fahr.
Nickel	790	1454
Element A	810	1490
Element B	825	1517
Element C	750	1382

The fourth element gave an indeterminate result, since it affects the two A_1 points, but does not decrease the carbon solubility.

The figures given above are for the Ac_1 point. The agreement seems sufficiently close, having in mind the circuitous experimental method enforced on us, to justify the belief that it is not entirely accidental.

Conclusions

The location of the A_3 stable point is so near that of A_2 in pure iron that one is led to consider the possibility that the beta phase may not exist in the stable system. The author has been unable to formulate a conception of a carbon-free stable system, and it is probable that the preceding suggestion were better expressed by saying that the stable A_3 point of the binary system approaches A_2 as the limit, when its carbon concentration is progressively decreased.

Honda[5] has shown that the Van't Hoff equation is applicable to the effect of carbon on the A_3 transformation. From the limitations of our method of attack, we lack data as to the effect of the first increments of carbon on the A_3 stable transformation point. If we postulate a straight A_3 stable line joining A_2 and the stable eutectoid point, and compare its slope at moderate concentrations with that of the established metastable A_3 line, we are forced to a conception that while austenite contains its carbon

[5] "On the Law of Depression of Freezing Points in Metallic Alloys," Kotaro Honda and Toyozo Ishigaki, *Science Reports* of the Tohoku Imperial University, Series I, Vol. XIV, No. 3, July, 1925.

in solution in the form of molecules, containing one atom of carbon each, boydenite, the stable solid solution, has as the solute, a molecule containing more than one, and probably three molecules.

It is realized that the problem is worthy of much more study, and that the conclusions offered may not yet be considered even as well founded hypotheses. Since, however, they have received confirmation from every angle, so far investigated, the author feels impelled to offer them here at least as a subject for discussion. It is regrettable that the data concerning the elements other than nickel and silicon are still to be regarded as confidential, and withheld from publication. There is hope that these facts may be released at some reasonably early date.

The author's thanks are extended to his several co-oworkers, Dr. A. F. Gorton, Dr. M. M. Austin, H. R. Payne, W. W. Flagle, G. M Guiler, H. H. Johnson, and Mrs. A. N. Hird, upon whose experimental observations this paper depends.

The Author

H. A. Schwartz was born in Oldham County, Kentucky, in 1880, graduated from Rose Polytechnic Institute in 1901, receiving the degree of Bachelor of Science; in 1903 he received the degree of Master of Science and the degree of Mechanical Engineer in 1905. Was instructor at Rose Polytechnic Institute in 1901 and 1902. From 1902 to 1920 at Indianapolis Works of the National Malleable Castings Company employed as draftsman, chemist, metallurgist, and assistant superintendent. In 1918 was metallurgical engineer for entire corporation, office at Indianapolis. From 1920 to date has been manager of research at the National Malleable and Steel Castings Company, Cleveland. Member of American Society for Steel Treating, being chairman of the Cleveland chapter, 1924-25 and 1925-26;

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DISCUSSION

Written Discussion: By Dr. Anson Hayes, Iowa State College, Ames, Iowa.

In the discussion of the paper "The Carbon Content of Pearlite in Iron-Carbon Alloys Containing One Per Cent Silicon[1]," which was read at the Cleveland convention the equation

$$(1) \qquad \frac{dT}{dX} = (K-1)\frac{RT^2}{\triangle H}$$

was given as representing the relation of the change in the temperature of the A_3 transition with change of mole fraction of a solute (carbon or silicon) when these are present in small mole fraction. In this equation T is the temperature of the transition. $\triangle H$ is the heat absorbed when one formula-weight of iron (55.84 grams) is changed from alpha (beta) to gamma, X is the mole fraction of the carbon or the silicon, as the case may be, and K is the distribution ratio of the carbon, or the silicon, between the alpha and gamma phases

$$\frac{\text{X in alpha}}{\text{X in gamma}}$$

It is generally conceded that carbon is essentially insoluble in alpha (or beta) iron and thus this equation takes the form

$$(2) \qquad \frac{-dT}{dX} = \frac{RT^2}{\triangle H}$$

as describing the influence of the first additions of carbon to pure iron. In other words, the temperature of the A_3 transition is lowered by addition of carbon. According to the report of Wever, given by Mr. Schwartz, the first additions of silicon raises the temperature of the A_3 transition. This is interpreted as meaning that K of equation (1) is greater than one. In other words, when the gamma and alpha phases are in equilibrium, the silicon distributes itself between the alpha and gamma iron in such a manner as to be present in the alpha phase in higher mole fraction than in the gamma phase.

[1] TRANSACTIONS, American Society for Steel Treating, Vol. 10, p. 214

In this paper[2] the fact was pointed out that if K for silicon is much greater than one a three-dimensional diagram will be required to represent the more exact behavior of the three-component system iron-carbon-silicon instead of the two-dimensional diagram which was proposed at that time.

The writer wishes to state that he considers the references given by Mr. Schwartz and Mr. Archer to be conclusive in showing that the temperature of A_3 is raised though it seems impossible at this time to determine how sharply the rise takes place with the first additions of silicon. It may also be that the two-dimension diagram will prove much more useful to the practical metallurgist due to its greater ease of interpretation than the more exact and more complicated three-dimensional diagram. It was due to these facts that no immediate change in the two-dimensional diagram there presented was made.

Since the presentation of the paper by Hayes and Wakefield, Kjerrman has shown in his paper "The Effect of Manganese, Silicon and Phosphorus upon the Pearlite Interval"[3] that the influence of a third component such as manganese, silicon or phosphorus is to spread the A_1 transition out over a temperature and composition interval. This spreading out of A_1 into an interval is the result of the unequal distribution of the third component between the austenite and the alpha iron and cementite which is forming. This spreading out found by Kjerrman is in complete conformity with the value of K for silicon in equation (1) being greater than unity.

Now Mr. Schwartz is attaching a significance to the rise in A_3 with silicon, which is entirely outside of anything that the writer has considered. Evidently, Mr. Schwartz's conclusions are arrived at from an entirely different method of approach than that which the present writer has used. It, therefore, seems best to present the accompanying diagram and the methods of arriving at it, as the most direct method of explaining our very different views of the significance of this rise in the A_3 transition with silicon.

This three-dimensional diagram is a modification of the earlier two-dimensional diagram and is changed so as to make it consistent with the assumption that at one per cent silicon the A_3 transformation in a carbon-free iron is raised to 1740 degrees Fahr. (950 degrees Cent.) This value is not considered as definitely established but it is the higher of the values given by Wever and Oberhoffer.[4]

The same general type of modification would result if A_3 rises by either greater or smaller value than the 120 degrees Fahr. (50 degrees Cent.) assumed here. The further assumption is made that at any composition to the left of the A_1 stable or A_1 metastable lines equation (1) gives the effect of silicon and equation (2) that of carbon upon the temperature of the A_3 transition.

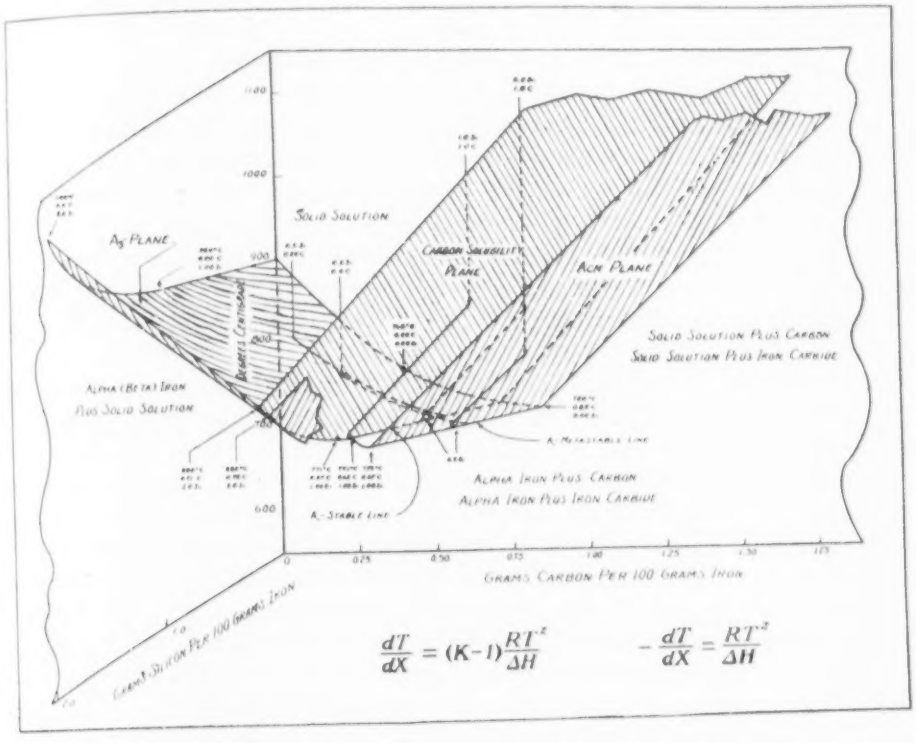
If pressure upon the three-component system iron-carbon-silicon be assumed to be constant, then the equation $P + F = 4$ gives $F = 2$ when alpha

[2] Loc. cit.

[3] TRANSACTIONS, American Society for Steel Treating, Vol. 9, p. 430

[4] Loc. cit.

iron and solid solution are in equilibrium, when carbon and solid solution are in equilibrium and when solid solution and cementite are in equilibrium. The A_2, the carbon solubility and the Acm lines become respectively the A_3, the carbon solubility and the Acm surfaces when alpha iron, carbon and solid



solution are in equilibrium $F = 1$ as it does also when alpha iron, cementite and solid solution are in equilibrium. This signifies, as Kjerrman has so nicely pointed out, that the A_1 point becomes A_1 lines i. e., in this three-component system, the A_1 stable and the A_1 metastable changes take place at a series of temperatures and compositions as represented by the A_1 stable and the A_1 metastable lines respectively.

In accordance with the foregoing treatment the A_2 surface is constructed from the following points

(Si = 0.00, C = 0.00, T = 900°C), (Si = 1.00, C = 0.00, T = 950°C) Wever
(Si = 0.00, C = 0.85, T = 720°C), To a point 2°C higher and very slightly to the left of the A_1 metastable (Si = 1.00, C = 0.67, T = 759°C) Hayes, Flanders and Moore
(Si = 0.00, C = 0.40, T = 768°C),

The Acm surface is drawn as a plane through

(Si = 0.00, C = 0.85, T = 720°C) and (Si = 1.00, C = 0.67, T = 759°C) and in a direction to intersect Si = 0.00, C = 2.00, T = 1134°C. Thus

the A_1 metastable line is the intersection of the A_2 and Acm surfaces.

This manner of drawing the A_2 surface makes it distinctly curved in the zero silicon plane and makes the surface very nearly a plane at a point slightly beyond the 1.00-per cent silicon plane. This was to conform to the fact that the so-called beta transformation is lowered from about 1415 to 1400 degrees

Fahr. (768 to 761 degrees Cent.) by the presence of this percentage of silicon.

In accordance with the method of constructing the two-dimensional diagram[5] the A_1 stable line in the three-dimensional diagram is considered to be the intersection of the carbon solubility plane with the A_3 plane.

We are not as fortunate in regard to data for locating the position of the A_1 stable line in space as we were in the location of the A_1 metastable, due to the fact that the temperature and carbon content of the iron-carbon eutectoid in the pure iron carbon system are not known. The point Si = 1.00, C = 0.57, T = 771°C should be very near the position of the A_1 stable line at 1.00 per cent silicon. This point is plotted on the accompanying diagram as being slightly below and to the left of the A_3 plane. The reason that it falls off of the A_3 plane is largely due to the fact that the value for T = 771°C was determined by Hayes, Flanders and Moore[6] upon a 0.95 per cent silicon alloy and the value 0.57 per cent carbon was taken from the solubility studies of Schwartz, Payne, Gorton and Austin[7] which work was done upon a 1.20 per cent silicon white iron.

A reasonable correction for the variation of carbon content accompanying a variation of silicon from 1.20 per cent to 0.95 per cent would easily place this point upon the A_3 surface.

NOTE—The writer wishes to state that the agreement between the carbon content of pearlite in a 1.00 per cent silicon white iron, as obtained by the method used in extending the A_3 line from the iron-carbon eutectoid to the temperature 1320 degrees Fahr. (771 degrees Cent.) as shown in the paper of Hayes and Wakefield[8] and that determined experimentally by Wakefield in this paper, is very good, indeed, if the above correction for the effect of variation of silicon upon the carbon content of the iron carbon eutectoid is made. This does not preclude, however, the possibility of the A_3 point being raised through a considerable interval by the presence of 1.00 per cent silicon as may be readily shown from the accompanying three dimensional figure.

The general direction of the carbon solubility surface with respect to temperature and carbon is fixed by the fact that in the 1.00 per cent silicon plane iron carbide is metastable and thus the carbon solubility plane lies to the left of the Acm plane. It is also rather closely fixed by the solubility studies of Schwartz, Payne, Gorton and Austin. The A_1 stable line was drawn parallel to the A_1 metastable line for want of other evidence than that of Maxwell and Hayes[9] which shows that iron carbide is metastable at 1330 degrees Fahr. (720 degrees Cent.) in the pure iron carbon system.

The behavior of a hypereutectoid alloy of about 1.00 per cent carbon and 0.5 per cent silicon when cooled slowly from a temperature above the critical range will be as follows: At the point where the 0.5 per cent silicon, 1.00 per cent carbon vertical line strikes the carbon solubility plane (see figure)

[5] Loc. cit.

[6] TRANSACTIONS, American Society for Steel Treating, Vol. 5, 1924, p. 183

[7] *Transactions*, American Institute of Mining and Metallurgical Engineers, Vol. 28, 1923, p. 212

[8] Loc. cit.

[9] Journal, American Chemical Society, Vol. 48, 1926, p. 584

temper carbon begins to separate. Since the separation of carbon will not greatly change the silicon content of the solid solution, the composition temperature values of the solid solution will follow the carbon solubility plane and will strike the A_1 stable line at a point very near the 0.5 per cent silicon point. From this point the separation of alpha iron will cause a reduction in the silicon content of the solid solution and this effect will be combined with that of the separation of carbon in such a manner as to cause the solid solution composition and temperature to follow the A_1 stable line in the direction of more carbon, less silicon and a lower temperature.

The behavior of a hypoeutectoid alloy upon cooling slowly from a temperature above the critical range and containing, for example, 0.5 per cent silicon and 0.25 per cent carbon would be as follows. At the point where the 0.5 per cent silicon, 0.25 per cent carbon vertical line intersects the A_3 surface (see diagram), alpha iron begins separating from the solid solution. The composition and temperature of the solid solution then follows the A_3 surface until the A_1 stable line is reached. Due to the fact that we have, in this discussion, assumed that silicon is more soluble in alpha iron than in gamma iron, the silicon content of the solid solution will be less than 0.5 per cent when the A_1 stable line is reached as shown in the figure.

From this point of intersection, if cooling is sufficiently slow, the composition of the solid solution and its temperature follows for a limited distance the A_1 stable line in the direction of lower temperature, more carbon and less silicon. This is because the alpha iron separating along the iron-carbon eutectoid line is reducing the silicon content of the solid solution. If the rate of cooling is too rapid in the case of either the hyper or hypoeutectoid alloy, the composition of the solid solution may reach the A_1 metastable line in which case the separation of alpha iron and iron carbide will take place in such a manner as to cause the compositions and temperature of the solid solution to follow the A_1 metastable line in the direction of less silicon, more carbon and a lower temperature. In case the cooling from the higher temperature is also sufficiently rapid, the Acm plane may be intersected in which case proeutectoid cementite will separate. This contingency is also shown in the figure.

The behavior accompanying the foregoing figure and discussion is the result of the writer's belief that austenite is essentially of the same fundamental nature whether it is formed from carbon or from iron carbide. Thus we may quote from a discussion of the paper of Hayes and Wakefield[10] by changing the words line to surface and point to line as follows. First, the fact that carbon may be taken into solid solution in austenite and precipitated from the solution as cementite and the reverse process that cementite may be taken into solution and the carbon may be precipitated indicates that there must be a mechanism for the conversion of carbon in each of these forms into the other. If such is the case, the solid solution is very probably the same in both cases. If we accept the concept of the solid solution as that which Wever presents and which reference Mr. Schwartz quotes, we may conclude that the carbon in the austenite enters the interior of the face-centered

[10]Loc. cit.

cube whether it comes from carbon or from cementite and that the iron atoms in the cementite take their place in the regular positions of the iron in these face-centered cubes. If this is correct, the cementite is formed at the time that the carbon is precipitated. At the iron-carbon eutectoid carbon is precipitated without the formation of the carbide. This being the case, it follows that the difference between austenite saturated with carbon from carbon and saturated with carbon from cementite is simply that a greater number of the face-centered cubes in the solution saturated with carbon from cementite have their interior occupied by a carbon atom. This concept at least conforms to our present knowledge of the structure of austenite as derived from X-ray studies. The writer has stated elsewhere that this same statement can be made of austenite in the presence of more than one form of the solute (dissolved carbon) if such forms are in equilibrium with each other.

In either of the above contingencies alpha iron should separate where the carbon solubility plane crosses the A_3 surface. For equilibrium processes there should be no discontinuity in the A_3 surface.

If these conclusions are correct, it becomes evident that there is only one A_3 surface. This being the case there is only one A_2 point in the pure iron-carbon system and it is the stable A_2 and occurs at about 1650 degrees Fahr. (900 degrees Cent.) Certainly there is no significance to the A_2 transition in pure iron except that it represents a stable transition for the phase gamma iron is a stable phase above its temperature and the alpha phase is a stable phase below it.

There is another fact which seems to the writer to be a fundamental objection to a value for A_3 stable in pure iron carbon alloys at any other temperature than the one which is usually taken 1650 degrees Fahr. (900 degrees Cent.). Suppose we assume that there are two molecular weights of the solute, one which we will call (C) and the other (Fe_3C). For dilute solutions the slope of the A_3 line is given by equation (2) and under such circumstances it may be integrated to the form $\triangle T = \frac{RT^2}{\triangle H} X$, where X is the mole fraction of either the (C) or (Fe_3C) as the case may be. Now in the iron-carbon diagram the abscissa is per cent carbon and at low values for carbon the value for X, if we assume (C) to be the solute, is very nearly the same as the value for X if we assume (Fe_3C) as the solute. The fact is that the values of X would not differ by more than 4 per cent even up to 0.5 per cent carbon. As zero carbon is approached they approach each other and at zero carbon become identical. Hence it appears that there would be only the one A_3 point at zero carbon and silicon even if it be granted that two absolutely independent solid solutions may exist. The writer, therefore, doubts the existence of an A_3 stable transition as distinct from the A_3 transition which we know either in pure iron-carbon alloys or in iron-carbon silicon alloys containing not more than one or two per cent silicon.

The writer regrets that there is not more dependable data from which the three-dimensional diagram might be extended, for example, a few more points for A_2, A_1 stable and A_1 metastable would be of great assistance in determin-

ing the manner in which these three change in curvature. It is of interest to note that according to Wever,[11] at silicon contents between 1.00 and 2.00 per cent temperature, A_3 transition in pure iron silicon alloys rises very rapidly. This makes the A_3 surface distinctly concave upward. Now if the Acm surface and the carbon solubility surface should remain nearly plane or should become concave downward the A_1 stable and A_3 metastable lines would both be turned upward and toward the right, that is, in the direction of less rapidly decreasing carbon contents and later to increasing carbon contents. If this represents the behavior, it is evident that neither the carbon content of A_1 stable nor A_1 metastable would become zero, and also it is evident that their temperatures would rise possibly above 1830 degrees Fahr. (1000 degrees Cent.) at higher silicon contents.

The foregoing discussion may thus be summarized as follows. The assumption that the A_3 transition is raised in temperature by additions of silicon combined with the assumption of the laws of dilute solutions for silicon and carbon in solution in gamma iron leads to a three-dimensional diagram which accounts for the rise in temperature of both A_1 stable and A_1 metastable as silicon increases. It also agrees qualitatively, at least, with the fact that the carbon content of the austenite at these two transitions first decrease and later increase as the silicon content increases.

According to this treatment there is only the one A_3 surface which is considered the A_3 stable transformation. It is further concluded that all methods of extrapolation for A_3 in the pure iron-carbon system by use of data obtained from a ternary system consisting of iron-carbon and a third element such as silicon or nickel should lead to the usual A_3 transition.

The writer wishes to acknowledge the meagerness of the actual experimental data available for the construction of the accompanying equilibrium diagram. Work is in progress in an attempt to obtain further critical data for its verification.

Written Discussion: By Oscar E. Harder, University of Minnesota, Minneapolis, Minn.

Definition of Terms: First, it must be definitely stated what is meant by A, A_2, and A_3, points or lines. In conformity to the well-established custom in the metastable equilibrium, the A, point on cooling for any given composition is the temperature at which the solid solution, austenite or Boydenite for that matter, becomes supersaturated with reference to iron and carbon and at which temperature these two phases separate out simultaneously. This point is called the Ar_1, or the eutectoid point. The above reactions are reversed on heating and the point is known as Ac_1. Obviously the A_1 point has no existence in pure iron and cannot be determined experimentally in the very low carbon steels.

While there is not general agreement regarding A_2, it is the writer's opinion that it is not a change of phase and therefore may be neglected in this discussion of the stable equilibrium.

The A_3 point is understood to mean the temperature at which the gamma to alpha transformation takes place on cooling and the alpha to gamma trans-

[11] Loc. cit.

formation on heating. This point coincides with the A_1 point, for the eutectoid reaction. The A_3 differs from the A_1, in that it does have a definite existence in pure iron.

Effect of Silicon on A_3 in Iron Silicon Alloys. From the information available at the present time, it seems rather well-established that as the silicon content increases, the gamma to alpha transformation is raised. It appears that the addition of silicon to iron lowers the delta to gamma transformation from about 2550 to about 2190 degrees Fahr. (1400-1200 degrees Cent.), at which latter temperature the line intersects the raised gamma to alpha transformation.

Composition Studied by Schwartz and Associates. The alloys which have been reported have contained in the neighborhood of 2.50 per cent carbon. Of course in the condition of hard iron these alloys do not represent the stable system. It is only after they have been annealed at a suitable temperature for a sufficiently long time to provide for the disappearance of cementite, probably by solution in austenite and precipitation as graphite or free carbon, that the stable equilibrium is established. The writer is aware of the fact that certain steels show a change from the metastable to the stable equilibrium at temperatures below the lower critical in the metastable system.

The alloys which have been studied by Mr. Schwartz and his associates have, therefore, always contained far more carbon than the eutectoid composition. Even though they may refer to silicon contents which reduce the carbon solubility to zero, it does not appear that they have studied any such alloys. Obviously in completely anealed malleable the phases are ferrite and carbon, but there is evidence for the opinion that ferrite is a solvent for carbon and that in reality such systems represent an equilibrium between a saturated solid solution of carbon in alpha iron and carbon.

Direction of $A_{1,2}$ Line: The author has taken two points determined in alloys of similar carbon contents, but of different silicon content, and then by drawing in a straight line which he extends to zero per cent of silicon postulated the position of A_3 stable. Such a procedure is surely open to question because he is assuming that the influence of the element silicon on the eutectoid point is an additive function of the amount of silicon in weight per cent. He has not limited his application to dilute solutions where such laws may hold with a fair degree of accuracy, but has used as one of the points an alloy containing 8 per cent silicon by weight. If he wishes to apply such a law it appears that it would be better to consider the molecular concentration.

Elements Reacting Similarly to Silicon: Little can be said upon this point because the elements are not named. A consideration of the crystal systems of the various metals might be interesting in this conection.

In conclusion it does not appear that there has been a sufficiently clear statement of terms, possibly there has not been the correct recognition of the active phases present in the stable system, and finally there has been an extrapolation which does not seem warranted. The writer would like to see the available data for the stable system iron-carbon-silicon worked up as a ternary system and finally made into models in which the vertical direction will

represent temperatures such as those of various critical points. Such a series of models was prepared by G. A. Rankin in his studies of the CaO,-Al_2O_3-SiO_2.

Author's Closure of Discussion

Dr. Hayes has given an instructive dissertation on the theory of a three-dimensional Fe-C-Si diagram. Assuming his postulates, there can be no question of the qualitative accuracy of his conclusions and he himself refers to the meagerness of quantitative data underlying his reasoning.

The difference in view point between Hayes and the writer is in that the former assumes a priori that "austenite is essentially of the same fundamental nature whether formed from carbon or iron carbide." While the author is not yet prepared to prove the contrary, he certainly is not prepared to admit the truth of this postulate. No reasoning based on Hayes' assumption, without proof of the latter's validity, can be logically adduced as evidence against the writer's view, for the proof then assumes the truth of a concept which the author submits to argument. Further, if the writer correctly interprets Dr. Hayes, the latter now assumes that carbon and silicon each affect the transformation point as though present alone, i. e., neither affects the effect of the other. Hayes seems previously to have categorically denied the validity of this assumption.

The suggestion that atom, or better, molecular concentration should form the basis of the argument, is fraught with difficulties, for we have as yet no conclusive evidence even as to whether elements are in solid solution as such or as chemical compounds. The author's paper, so far as he is aware, offers the first suggestion that carbon may possibly be combined under these circumstances into a molecule containing over one atom of carbon. It is not believed that real proof of this concept is anywhere available. The work of Yamada as quoted by Honda (loc. cit.) would lead to the belief that solid solutions of the Fe-Fe_3C systems are "dilute" only when they contain but few hundredths of 1 per cent carbon. Any rectilinear relations are probably empirical in any event. The extrapolated value of Si for zero carbon to which Dr. Hayes points, finds some support from the data of Becker. "Equilibrium at High Temperatures in the Iron-Carbon-Silicon System."[12] The author does not necessarily endorse all of Becker's views but the data point strongly to the insolubility of carbon in the presence of about 4 per cent silicon.

If we assume that the position of a solvent power for carbon is proof of the presence of iron in the gamma phase, then the solubility of carbon at any temperature for silicon contents greater than that at which the A_3 and A_4 points merge, is evidence in favor of a difference between the A_3 points of the two systems. Here again we may refer to Becker for proof that carbon is quite soluble at high temperatures in the presence of 4 per cent silicon. Unless the addition of carbon to the Fe-Si system increases the ordinate of the merging of A_3 and A_4 rapidly, a point on which we have no evidence of any kind, we must postulate a difference in the locus of the A_3 surfaces of the two systems.

The author has no wish to insist too strongly upon the locus of A_3 stable

[12] *Journal* of the Iron and Steel Institute, 1925, No. II, Vol. CXII, p. 249

with decreasing carbon and silicon (or other graphitization accelerator). It would seem, however, very surprising that all the work, with five different elements, should have been affected by a series of accidental errors of such character as to affect the result so nearly equally. He would be much gratified if some one could show a reason, different from that suggested, why the results should agree so nearly.

The fundamental observations are the effect of the graphitizing element on A_1 stable and metastable and upon the eutectoid carbon ratio. Beyond small concentrations these become nearly rectilinear. Is there any plausible source of error affecting all three determinations systematically so that irrespective of the system chosen if $\triangle T_s$ and $\triangle T_m$ are the changes in the stable and metastable A_1 points for unit addition of the graphitizing element and X be the concentration of that element reducing the eutectoid carbon concentration to O, X ($\triangle T_s$-$\triangle T_m$) will necessarily be nearly constant? If so, the author's argument is valueless. If not, it points either to a surprising coincidence or else establishes at least a tentative conclusion as to the locus of A_1 in the hypoeutectoid stable system.

THE DECOMPOSITION OF THE AUSTENITIC STRUCTURE IN STEELS—Part IV*

Effect of Stress on the Decomposition of Austenite

By Ralph L. Dowdell and Oscar E. Harder

Abstract

This article reports a series of tests which was made to determine the effect of stress on the stability of austenitic structures. An attempt has been made to distinguish between the effect of compression and tensile stresses. The steels included in the investigation are the same as those described in Part I of this paper. The experiments have included tests on the effect of bending stresses during tempering, severe upsetting at room temperature and during tempering, tensile stresses at room temperature and at the temperature of carbon dioxide snow. It has not been possible to produce martensite by bending stresses or by severe upsetting. Typical martensitic needles have been produced by tensile stresses in ring specimens and by subjecting specimens which had been stressed in tension to the temperature of carbon dioxide snow. Stresses have been found to make austenitic structures less stable both at low and at elevated temperatures.

Effect of Stress on the Tempering Changes

IT was believed that tensile stress would promote the decomposition of austenite on tempering and that compressive stress would retard it because when austenite decomposes into either troostite or martensite there must be an increase in volume.

In order to test out this theory an apparatus was designed to stress certain areas in tension and other areas in compression at the same time, and to subject the specimen at various stresses to the tempering treatments.

Description of Stressing Apparatus—The stressing apparatus consists of a high carbon steel yoke which supports the test piece

*This and five other articles on The Decomposition of the Austenitic Structure in Steel are based on a manuscript submitted by Ralph Lewis Dowdell in partial fulfillment of the requirements of the Graduate School of the University of Minnesota for the degree of Doctor of Philosophy. O. E. Harder, in charge of research. These articles have appeared serially in Transactions since January, 1927. Seven installments constitute this series.

A paper presented before the eighth annual convention of the Society, Chicago, September 20 to 24, 1926. Of the authors, Dr. R. L. Dowdell is assistant professor of metallography and Dr. O. E. Harder is professor of metallography, University of Minnesota, Minneapolis.

at both ends as shown in Fig. 113. The stress is applied to the bar by screwing up the 3/8-inch bolt with a wrench. The copper dial of the apparatus is graduated in degrees. On the calibration of the apparatus with an Ames dial it was found that if the stressing bolt is turned through an angle of one degree, the screw would travel a distance of 0.000342-inch.

With a transverse stressing of this kind, the side of the bar opposite the screw is in tension while the side of the bar in contact with the screw is in compression. Boyd[1] gives the following formula for the deflection of bars supported at both ends and loaded in the middle:

$$f = \frac{1}{48} \times \frac{Pl^3}{EI}$$

P = Load in pounds
l = Length in inches between supports (2.75")
E = Modulus of elasticity (about 27,000,000 for austenite)
I = Moment of inertia =

$$\frac{1}{64} \pi d^4$$

f = Deflection at the center in inches (0.175")
M = Bending moment = ¼ Pl
c = ½ the diameter in inches
S = Stress in extreme outer fibres in pounds per square inch

$$S = \frac{Mc}{I} = \frac{Plc}{4I}$$

$$\text{or } P = \frac{4SI}{lc}$$

$$f = \frac{4SI\,l^2}{48EIlc} = \frac{Sl^2}{12Ec}$$

$$\text{or } S = \frac{12fEc}{l^2}$$

This last formula was used to calculate the outer fiber stress in the bars (0.35 inch diameter x 2.75 inches long between supports) before the whole apparatus was placed into the tempering oven. In this calculation $\frac{3}{10}$ of the stress as calculated from the formula was deducted in order to compensate for the deflection at the center of the yoke. Previously the apparatus had been

[1]Text Book, Strength of Materials, page 162. McGraw-Hill Co.

seasoned in the oven at 392 degrees Fahr. (200 degrees Cent.). About the only change that could be noticed in the apparatus was the slight oxidation of the copper dial.

Tempering Under Stress—Table IX gives the calculated stress, tempering temperature and time for steels Nos. 3, 6, 8 and 10. Microscopic examination does not show any appreciable structural

Table IX

Data of Experiments on Tempering Under Stress

Steel	Size Diam. inches	Quench Temp.	Time at Temp. Minutes	Temper in days at °F.	Instrument Reading in Degrees	Calculated stress on outer fiber	Remarks
3	0.35	1150° C. (2102° F.)	15	8 at 392	14	24,000	Released by hand; permanent set.
6	0.35	1150° C.	15	1 at 392	20	35,000	As above. Bar cracked on stressing (T. side).
8	0.35	1150° C.	15	2 at 392	25	43,000	Released by hand; perm. set. Bar cracked on stressing (T. side).
8	0.35	1040° C. (1904° F.)	30	not tempered	70	124,000	Broke on stressing. Not tempered.
8	0.35	1040° C.	30	not tempered	50	88,000	Broke on stressing. Not tempered.
10	0.35	1100° C. (2012° F.)	30	4 at 212	20	35,000	Released by wrench 20 degrees. Bar showed no cracks.

NOTE: All bars quenched from carbon pack in oil at room temperature.

difference between the tension and compression sides. All of these steels show martensite on the outside after quenching (Figs. 115 and 117). Nos. 3 and 10, however, were only partly martensitic. It, therefore, resulted in these tests that the martensite was tempered to troostite. It now appears that the size of the test specimen was not well chosen and better results may be expected with larger sections.

After the flexure stress was first applied to the bars, they were released and were found to have returned to their original position. However, these stress tests, on tempering showed that a permanent set was obtained after tempering under stress below the limit of proportionality at a temperature under 392 degrees Fahr. (200 degrees Cent.).

Effect of Mechanical Deformation on the Decomposition of Austenite at Room Temperature

The following experiments were carried out in order to study

the effect of mechanical deformation on the decomposition of austenite at room temperature.

Austenite at this temperature is in a metastable condition and it is known that agitation, mechanical working, etc., will serve as a means of bringing about the transformation of a metastable to a stable equilibrium. Two types of tests have been included: one in which the specimens have been compressed by hammering and the other in which they have been placed in tension.

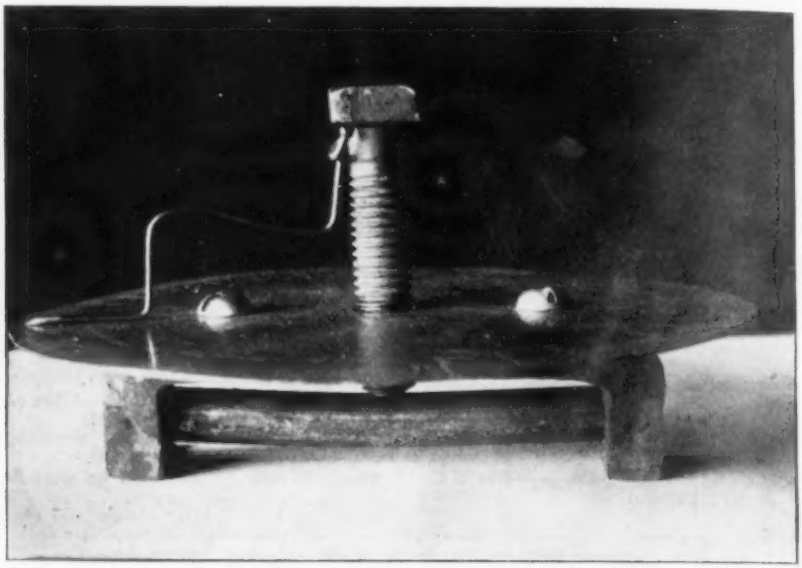
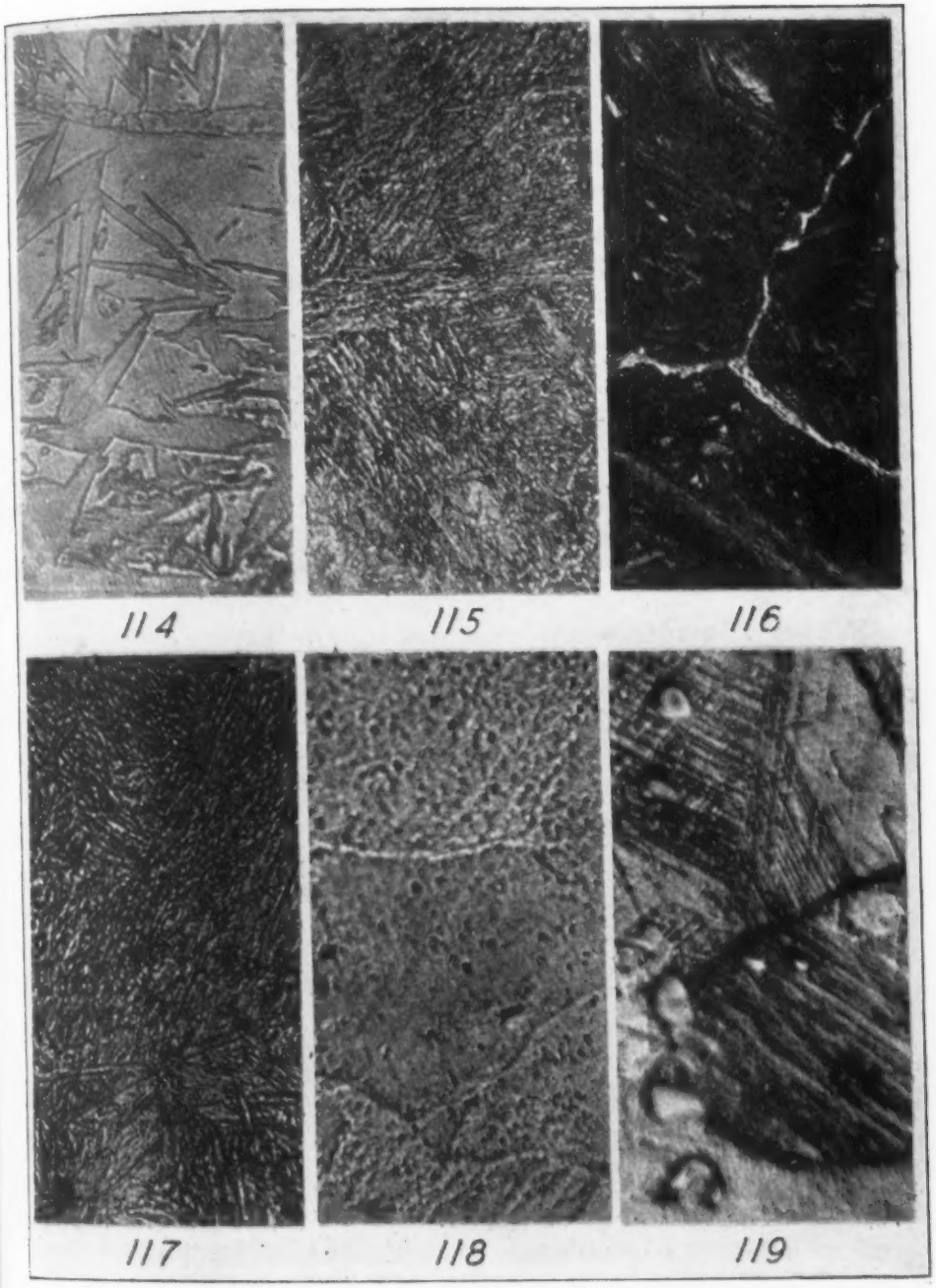


Fig. 113—Photograph of Stressing Apparatus.

Effect of Upsetting by Hammering—Table X shows the effect of cold deformation on the scleroscope hardness of several austenites. Some of the specimens were hammered on an anvil with a hammer weighing about two pounds and others were upset with a steam hammer.

Photomicrographs Figs. 119 and 120 show only the customary slip lines having the characteristic angles within any one grain but otherwise they do not resemble martensite. However, it should not really be expected that these complicated compressive stresses would aid in the recrystallization to the typical martensite but these stresses should nevertheless aid in recrystallization, both at room and at elevated temperatures. As previously reported, a specimen of steel No. 3 after upsetting was tempered



Decomposition of Austenite and Martensite on Tempering Under Stress. Steel No. 8. Quenching Treatment: 2102 Degrees Fahr., Oil Tempered Two Days at 392 Degrees Fahr. Calculated Outer Fiber Stress 43,000 Pounds Per Square Inch. Fig. 114—Center of Bar as Quenched. Etched with Picric Acid. Austenite ± Martensite + Carbide. Mag. 1000x. Fig. 115—Edge of Bar as Quenched. Etched with Picric Acid. Martensite. Mag. 1000x. Fig. 116—Center on Neutral Axis After Temper. Etched with Picric Acid. Austenite + Troostite + Carbide. Mag. 1000x. Fig. 117—Outside Edge After Temper Same on Compression as on Tension Side. Etched with Picric Acid. Troostite. Mag. 1000x. Effect of Mechanical Stress on the Decomposition of Austenite at Room Temperature. Complex Stresses Produced by Hammering. Fig. 118—Electrolytic Iron After Upsetting With Power Hammer. Deeply Etched with Picric Acid. Ferrite + Etching Pits. Mag. 1000x. Fig. 119—Manganese Steel No. 2 After Upsetting With About No. 2 Hammer. Etched with Picric Acid. Austenite + Octahedral Slips. Mag. 1000x.

at 374 degrees Fahr. (190 degrees Cent.) for 4 days and it was found that it had tempered completely to troostite (Fig. 85) while an unstrained specimen of the same steel showed less than half of the austenite to have been decomposed in 4 days at 392 degrees Fahr. (200 degrees Cent.) (Fig. 81).

Tensional Deformation of Ring Specimens—It appeared that in order to produce a typical martensite from austenite at room temperature it would be necessary to make a specimen in the form of a ring so that it could be expanded under mostly tensional stress.

Table X

Effect of Cold Deformation by Hammering on the Scleroscope Hardness of Austenite

Steel No.	Hammer	Size in inches before hammering	Size in inches after hammering	S. h. n. before—after hammering
Elec. Fe	steam	1/2 diam. x 1/2 high	3/4 diam. 1/8 high	31.9 41.9
1	2 lb.	3/8 diam. x 3/8 high	3/16 high	46.0 64.0
2	2 lb.	3/8 cube	3/16 high	30.1 74.1
3	2 lb.	.375 cube	3/16 high	46.4 62.3
3	steam	.375 cube	1/4 high 1/2 square	47.2 85.1
4	2 lb.	3/8 cube	3/16 high	53.6 63.2

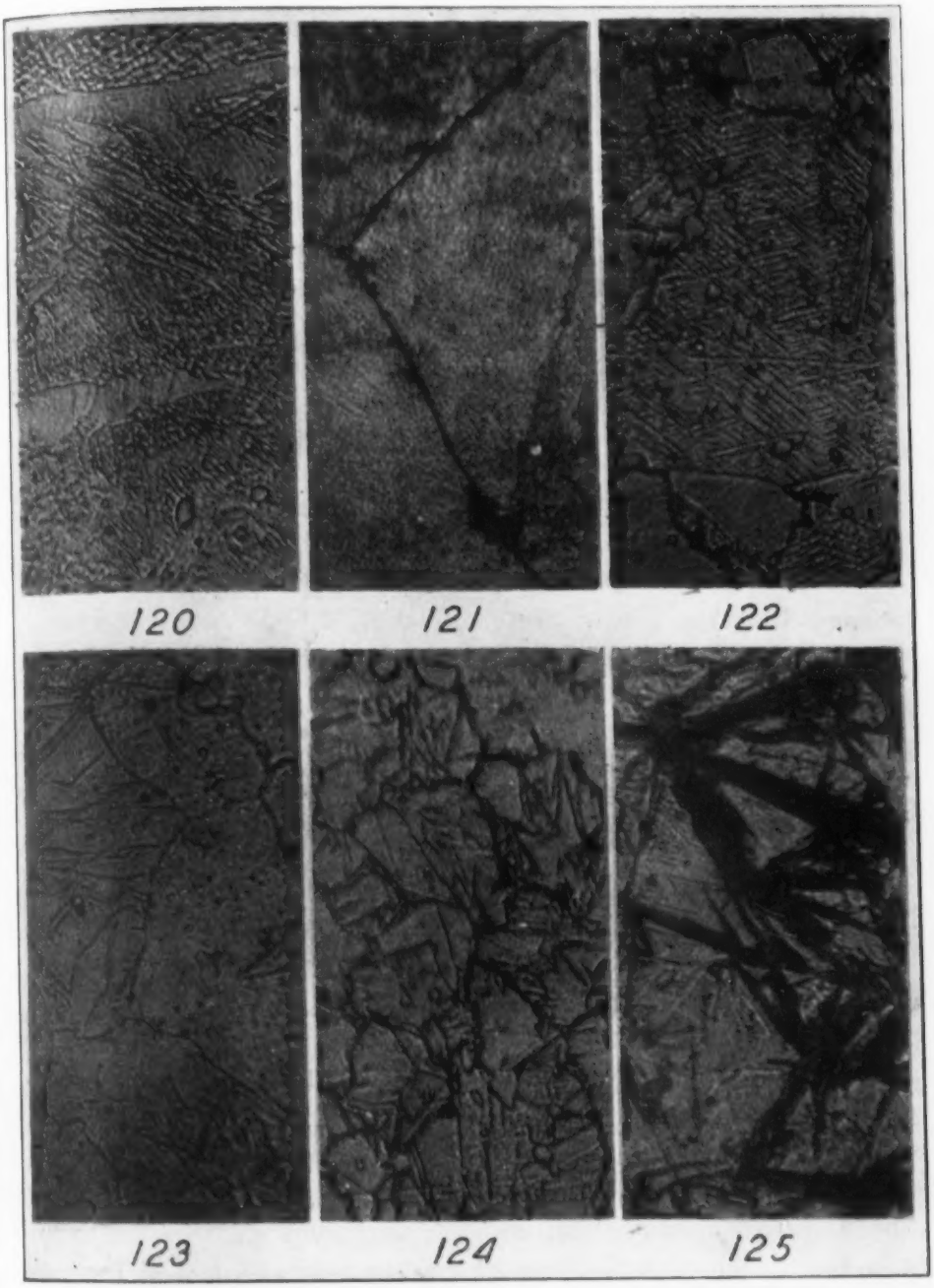
Microstructure showed slip lines only in all specimens.

Four ring specimens were machined, two of steel No. 3 and two of steel No. 8. These rings had an inside diameter of $\frac{7}{16}$ inch, a wall thickness of $\frac{1}{8}$ inch and a width of $\frac{1}{4}$ inch. Ring No. 2 of steel No. 3 however had a wall thickness of only $\frac{1}{16}$ inch.

Before quenching, these rings were taken one at a time and tightened between washers on $\frac{3}{8}$-inch bolts through their centers. They were then plastered with alundum cement to a thickness of about $\frac{1}{4}$ inch so as to retard quenching and help eliminate the martensite that would form on the edges of these small sections. They were then heated for quenching under charcoal and oil-quenched as shown in Table III and in the following discussion.

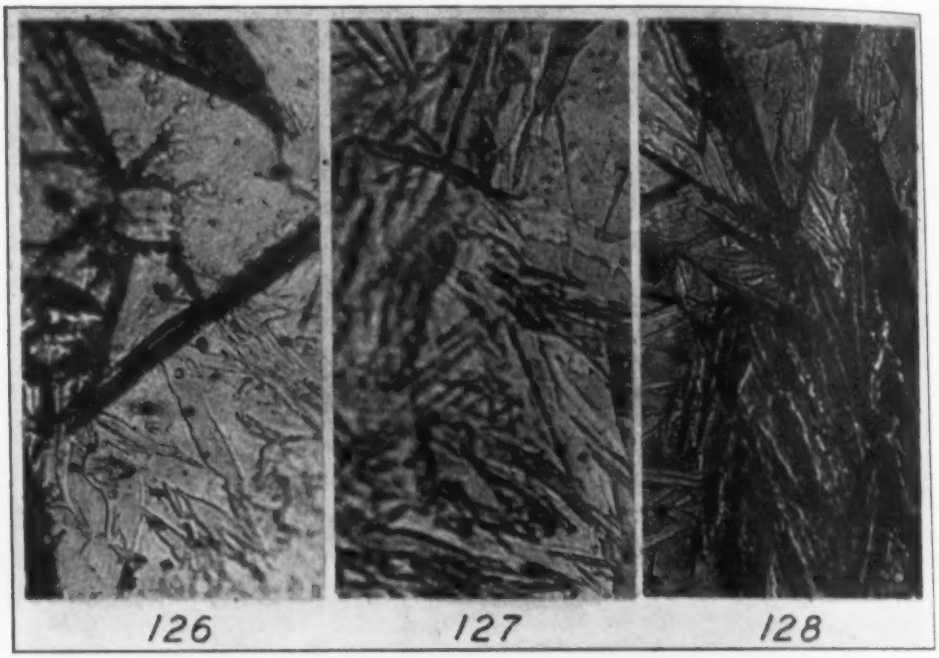
The stressing of the rings was obtained by forcing a tapered arbor (taper of 0.01 inch per inch of length) into the rings. In this operation the rings were placed against the end of a short piece of pipe and the taper gradually forced in by tightening the screw of a vise.

Rings Nos. 1 and 2 of steel No. 3 were austenitic to the edges but they contained some carbide after the quenching. Ring No.



Effect of Mechanical Stress on the Decomposition of Austenite at Room Temperature. Fig. 120—Steel No. 1 After Upsetting with a No. 2 Hammer. Mag. 1000x. Effect of Tensile Stress at Room Temperature on the Decomposition of Austenite. Steel No. 3. Fig. 121—Ring No. 1 After Quench. Mag. 1000x. Oil-Quenched After 30 Minutes at 2102 Degrees Fahr. After Stressing to a Calculated Stress of 30,000 Pounds Per Square Inch the Specimen Broke. Fig. 122—Ring No. 2 After Stressing to Failure. S. h. n. = 48.0. Mag. 1000x. Oil-Quenched After 15 Minutes at 2066 Degrees Fahr. Steel No. 3. Figs. 123 and 124—Ring No. 2. S. h. n. = 59.2. Mag. 1000x. Effect of Tensile Stress at Room Temperature on the Decomposition of Austenite. After Stressing, Ring No. 2 was Held About 4 Hours in CO_2 Snow and Allowed to Come to Room Temperature Over Night. Steel No. 8. Ring No. 4. Quenched in Oil From 2282 Degrees Fahr. After 30 Minutes; Tempered 8 Hours at 320 Degrees Fahr. Stressed by Taper from 0.7000 Inch Outside Diameter to 0.7005 Inch. Micrographs All on Extreme Outside Edge. Fig. 125—After 320 Degrees Fahr. Temper. White Needles Present Only at this Place. Mag. 1000x.

1 had such a coarse grain after its heat treatmnet [30 minutes at 2100 degrees Fahr. or 1150 degrees Cent.] that it broke at a calculated stress of only about 30,000 pounds per square inch. This value, of course, is only approximate because the ring may have had some initial stress and in the calculation the modulus of elasticity of the austenite is not definitely known. Ring No. 2



Effect of Tensile Stress at Room Temperature on the Decomposition of Austenite. Steel No. 8. Ring No. 4. Quenched in Oil from 2282 Degrees Fahr. After 30 Minutes; Tempered 8 Hours at 320 Degrees Fahr. Stressed by Taper from 0.7000 Inch Outside Diameter to 0.7005 Inch. Photomicrographs All on Extreme Outside Edge. Fig. 126—After Stressing Specimen Showed Considerable Amount of White Needles Only on Outside Edge. Mag. 1000x. Fig. 127—After Stressing. Mag. 1000x. Fig. 128—After Stressing. Mag. 1000x.

showed a much finer grain owing to its heat treatment [15 minutes at 2066 degrees Fahr. (1130 degrees Cent.)] but after stressing showed many slip lines and not the typical martensitic needles (Fig. 122). After several weeks at room temperature this ring (No. 2) was cooled in CO_2 snow (about —57 degrees Cent.) for four hours. On examination it was found that considerable recrystallization had taken place and characteristic martensitic needles were developed (Figs. 123 and 124). The scleroscope hardness after stressing was about 48 and the piece was practically nonmagnetic while after the CO_2 treatment the hardness in the vicinity of the break was about 59 and the piece was quite strongly magnetic. During this experiment a piece of steel No. 3 of austen-

itic structure and also a piece of quenched steel No. 10 mostly austenitic but having many carbide needles, were cooled and it was found that the CO_2 immersion produced no change in their microstructure. This seemed to indicate that previous tensional stressing aided in the recrystallization to the typical martensite.

In the quenching of rings 3 and 4 of steel No. 8, it was found that it was quite impossible to eliminate some of the martensitic formation. After quenching, these rings were tempered about 8 hours at 320 degrees Fahr. (160 degrees Cent.) so as to decompose the martensite, otherwise it would have been impossible to identify any new martensite formed. Fig. 125, however, shows several white needles at the edge of the specimen after tempering and the only ones on the section. Immediately after stressing there were quite a large number of the typical martensite needles produced on the outer edge of the ring, some of which are shown in Figs. 126 to 128. After retempering this ring at 320 degrees Fahr. (160 degrees Cent.) for about 8 hours all of these needles were darkened and no further needles were produced.

This test shows quite conclusively that martensite needles can be produced mechanically from some austenites by simply applying tensile stress.

Results and Conclusions

The experiments on the decomposition of austenite by mechanical stresses at various temperatures lead to the following conclusions:

1. When austenitic-martensitic steels are tempered under transverse stress below their so-called yield point at temperatures between 212 and 392 degrees Fahr. (100 to 200 degrees Cent.), a permanent set results. The bar is longer on the tension side which indicates that tensional stress favors the change from the gamma to alpha state. Microscopic examination, however, showed no difference in structure on the tension and compression sides. It is believed that more extensive tests of this type on larger bars may furnish additional information regarding this mechanism.

2. It has not been possible to produce martensite by the complicated stresses set up in austenite when it is upset at room temperature. Prominent slip lines have been produced, but they are characteristically different from martensite needles.

3. Characteristic martensite has been produced from austen-

ite at room temperature by tensile stress. It is very probable that all martensite is formed by a similar process in the normal hardening of tool steels but at a much more sudden and higher tensional stress.

4. An austenitic specimen which had been previously deformed in tension was caused to transform to martensite at about —70 degrees Fahr. (—57 degrees Cent.) while unstressed specimens were unchanged by this same treatment.

5. An austenitic specimen which had been deformed by hammering was found to decompose more readily on heating than an unstressed specimen.

6. The general conclusion may be drawn that deformed austenite is rendered less stable on cooling or on heating.

Educational Section

These Articles Have Been Selected Primarily For Their Educational And Informational Character As Distinguished From Reports Of Investigations And Research

FACTS AND PRINCIPLES CONCERNING STEEL AND HEAT TREATMENT—Part XII[1]

H. B. KNOWLTON

Abstract

This article covers the composition, properties, uses and heat treatments of some of the common types of chromium-nickel steels. The low, medium and high chromium-nickel steels with carbon contents from 0.10-0.55 per cent are discussed. Some of the nickel-chromium heat resisting alloys are also discussed.

NICKEL-CHROMIUM STEELS

AND

NICKEL-CHROMIUM AND HEAT RESISTING ALLOYS

IN the previous discussion of alloy steels, the effects of chromium and nickel were considered separately. It was stated that nickel forms an alloy with the ferrite (or iron) but not with the carbon. Thus it strengthens and toughens the ferrite and consequently strengthens and toughens the steel. Chromium, on the other hand, forms a double carbide with the iron and carbon in the steel. Thus the cementite of a chromium steel contains chromium carbide as well as iron carbide. Due to the properties of this chromium carbide, the addition of chromium to steel increases the hardening power without a proportionate reduction in the toughness. Steels containing small amounts of chromium are not only harder than the corresponding plain carbon steel but they may be hardened with a slower speed of quenching.

By the addition of both chromium and nickel a series of

[1]This is the twefth installment of this series of articles by H. B. Knowlton. The several installments which have already appeared in TRANSACTIONS are as follows: March, June and October, 1925; January, April, May, June, August, October, December, 1926; March, 1927.

The author, H. B. Knowlton, member of the Fort Wayne Group of the Society, is metallurgist of the Fort Wayne Works, International Harvester Company, Fort Wayne, Ind.

steels may be produced which show the beneficial effects of both of these alloying elements. For this reason there is on the market today a group of steels known as chromium-nickel steels which are used in automotive construction and for other parts which must have increased strength, toughness and hardening power. The common types of chromium-nickel steel contain at most only a few per cent of the alloying elements. Consequently they fall in the class of pearlitic steels according to Guillet's scheme of classification. They are capable of being annealed, hardened, tempered and heat treated much the same as the corresponding plain carbon steels. It is true that the heat treating temperatures and the method of quenching may be different, but the principles are the same.

The common types of chromium-nickel steels contain from 0.10-0.55 per cent carbon; 1.0-3.75 per cent nickel; and 0.45-1.75 per cent chromium. While other combinations may be possible these are the most common. The specifications of the Society of Automotive Engineers describe four groups of chromium-nickel steels. With a few exceptions these steels have the following chemical composition.

Chemical Composition of the S. A. E. Chromium-Nickel Steels

	3100 Series	3200 Series	3300 Series	3400 Series
	Per Cent	Per Cent	Per Cent	Per Cent
Manganese	0.50-0.80	0.30-0.60	0.30-0.60	0.45-0.75
Phosphorous	0.04 max.	0.04 max.	0.04 max.	0.04 max.
Sulphur	0.045 max.	0.04 max.	0.04 max.	0.04 max.
Nickel	1.00-1.50	1.50-2.00	3.25-3.75	2.75-3.25
Chromium	0.45-0.75	0.90-1.25	1.25-1.75	0.60-0.95

The manganese content of S. A. E. 3115 and 3120 is specified as 0.30-0.60 per cent. This is in accord with the general practice of holding the manganese content of case hardening steels comparatively low, except in cases where the manganese is used as an alloying element. As stated before the carbon content of the above steels varies from 0.10-0.55 per cent depending upon the purpose for which the steel is to be used. As a general rule the specifications allow a 0.10 per cent variation in the carbon content. Thus 3115 means a steel of the 3100 series with a carbon content of 0.10-0.20 per cent, while 3135 is a similar steel with 0.30-0.40 per cent carbon. Some of the users of alloy steels have

found that it is advisable to specify the carbon content within 0.05 per cent variation for steels used for certain particular jobs. When such specifications are imposed the consumer necessarily pays a higher price for the steel.

The principal difference between the four classes of chromium-nickel steel is in the content of chromium and of nickel. As might be expected the effect of the alloying elements upon the properties of the steel increases as the content of the alloying elements increases. On the other hand the higher alloy steels are more expensive in proportion to the amount of the alloying elements used. Also the higher alloy steels may be somewhat more difficult to forge and machine. Consequently the lower alloy steels are used wherever they will give satisfactory results. For parts demanding exceptional strength and toughness the higher alloy steels must be resorted to. The ratio of chromium to nickel is not exactly the same in the four steels described above. The nickel content is about 1.5-4.0 times that of the chromium. Some writers state that the nickel content should be about 2.5 times that of the chromium.

Among the early chromium-nickel steels were the Mayari steels. These were made from ores found at Mayari, Cuba, which contained both chromium and nickel. According to Camp and Francis, the steels made from these ores usually contain 1.0-1.5 per cent nickel and 0.20-0.70 per cent chromium. Making a steel from an ore which contains both of the alloying elements has obvious advantages. It is, however, more difficult to produce a uniform product in this manner. It is generally conceded that a better and more uniform steel may be made synthetically.

The addition of nickel to steel lowers the critical point and consequently the hardening temperature, whereas the addition of chromium to steels will require higher quenching temperatures than those employed for the corresponding plain carbon steels, due to the greater difficulty of dissolving the chromium carbides. When chromium and nickel are both present in steel in the proper proportions, the quenching temperatures may be approximately the same as those employed for the corresponding plain carbon steels. In general the chromium-nickel steels will harden with a slower speed of cooling than the corresponding plain carbon steels. This means that frequently oil quenching may be substituted for

water quenching, also that there will be a deeper penetration of the hardening effect.

Properties and Uses of Certain Chromium-Nickel Steels

Case Hardening Steels—The low carbon chromium-nickel steels are used primarily for case hardening although with suitable heat treatments they may also be used for some structural purposes. These steels have a distinct advantage over the plain carbon steels for case hardening since with their use case hardened parts can be produced which have greater strength and toughness and greater surface hardness. With the use of chromium-nickel case hardening steels the trouble with soft spots which is so common in case hardening may be greatly reduced. Also when chromium-nickel steels are used it is frequently possible to produce sufficient hardness by giving the final quench in oil. This means less trouble with warping and distortion.

S. A. E. 3115 and 3120 Steels—The S. A. E. 3115 and 3120 steels are the low chromium-nickel steels with the proper carbon content for case hardening. They are probably used more frequently for this purpose than are the higher alloy steels. The S. A. E. 3115 steel is recommended for most case hardened parts because the low carbon content (0.10-0.20 per cent) assures a tough core for two reasons: first because the lower the carbon content the less will be the hardening power; second, because the greater the difference between the carbon content of the core and the case the greater will be the difference between their critical points. When there is considerable difference between the critical points of the core and the case, it is possible to give the final quench from a temperature which is high enough to harden the case without producing much hardening effect upon the core. When the critical points of the core and the case are close together, a quench which is high enough for the case will have considerable effect upon the core. If the operation is not carefully performed and the temperature of the quench is a little higher than necessary, the core may become somewhat hard and brittle all the way through. For these reasons it is customary to hold the carbon content below 0.20 per cent for most parts. S. A. E. 3120 steel is recommended only for massive parts and then a caution is given to perform the heat treatment with great care to avoid producing brittleness of the core. The reason for using

the slightly higher carbon content (0.15-0.25 per cent) for large massive parts is that such parts naturally cool more slowly than do small ones. Consequently the slightly higher carbon content produces about the same physical properties in the core as are produced by the lower carbon content with the faster quench received by small pieces.

S. A. E. 3215 and 3220 Steels—These steels are similar to the S. A. E. 3115 and 3120 steels except that the content of the nickel and chromium are higher. They are sometimes described as medium chromium-nickel steels. The Society of Automotive Engineers recommends their use particularly for parts having sections more than 1.5 inches thick. In general, it may be said that they should be used for case hardened parts requiring greater strength and toughness than can be produced by the use of S. A. E. 3115 and 3120 steels. As mentioned in the case of the low chromium-nickel steels, the 0.15 per cent carbon steel is recommended for most purposes.

S. A. E. 3312 and 3415 Steels—These are high chromium-nickel case hardening steels. They are recommended only for parts requiring exceptional strength and toughness.

Medium Carbon Chromium-Nickel Steels

S. A. E. 3125, 3130, 3135, and 3140 Steels—These are low chromium-nickel steels of medium carbon content which are much used for parts requiring strength and toughness. A great many of the forgings used in automotive construction are made from these steels. These grades of steel may also be purchased in the form of bar stock for machining into parts which are sufficiently uniform in shape that they do not require forging. The lower carbon grades are used for parts which primarily require great toughness, while the S. A. E. 3135 and 3140 are used for those parts which require greater strength. Among the parts made from these steels, may be mentioned, axles, steering knuckles, connecting rods, bolts, etc. A heat treatment is necessary in order to bring out the best properties. Bar stock if properly finished in the mill may not need normalizing but all forgings should be normalized to break down the coarse forge structure before proceeding with other heat treatments. The heat treatment consists in heating to above the upper critical point followed by quenching either in water or oil and then tempering until the required

physical properties are produced. The S. A. E. 3125 and 3130 steels are usually quenched in water while the higher carbon chromium-nickel steels are usually oil-quenched.

For definite information concerning the physical properties which may be produced by various heat treatments applied to these steels and to the other chromium-nickel steels, the reader is referred to the American Society for Steel Treating HANDBOOK, and the data sheets put out by the International Nickel Company

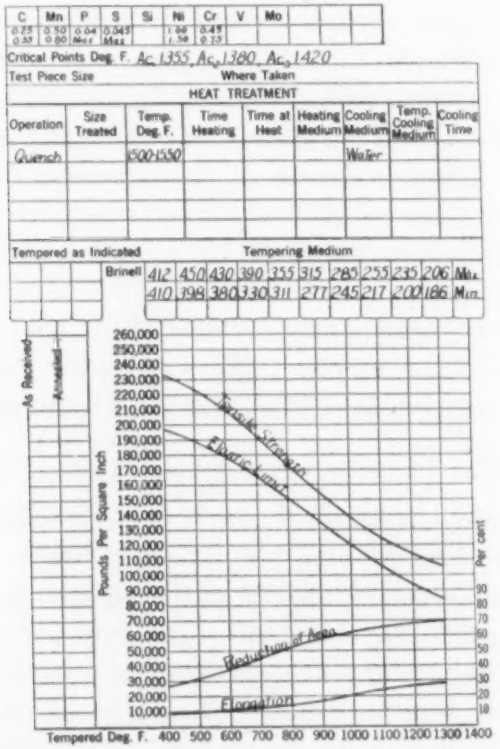


Fig. 1—Physical Properties Chart of a Nickel-Chromium S. A. E. 2130 Steel. Quenched in Water and Tempered.

and the Society of Automotive Engineers. Two charts are given herein showing the physical properties of S. A. E. 3130 steel after water quenching and S. A. E. 3145 steel after oil quenching followed in each case by tempering to different temperatures. The physical properties produced by similar treatments applied to S. A. E. 3135 and 3140 steels will fall in between those given on the two charts. In general the higher the carbon content and the faster the quench, the higher the tensile strength, elastic limit

(or yield point) and the hardness number will be, and the lower will be the elongation and the reduction of area.

It will be noted from these charts that these steels have considerable hardening power. Consequently they may be used for parts requiring a moderate resistance to wear. For parts requir-

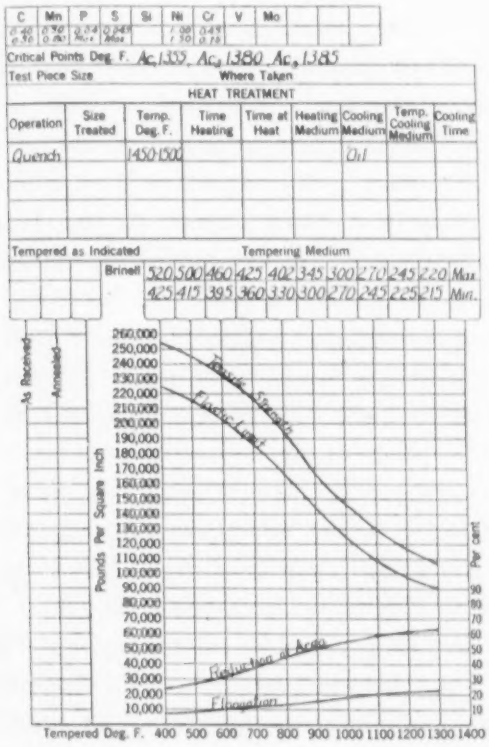


Fig. 2—Physical Properties Chart of a Nickel-Chromium S. A. E. 3145 Steel. Quenched in Oil and Tempered.

ing extreme hardness or resistance to wear it is necessary to use either a high carbon steel or resort to case hardening a lower carbon steel. The corresponding chromium-nickel steels with a higher content of chromium and nickel also have increased hardness under the same treatments.

Whether these steels can be machined after heat treating depends upon the tempering temperature. If high tensile strength is required and a low tempering temperature is employed, it may be impracticable to machine after heat treating. In such cases the article must be finish machined, with the exception of grinding operations, before heat treating. On the other hand there are many parts which are given high tempering treatments in order

to produce the required toughness which are soft enough after heat treating to be machined. It should be mentioned that machinability does not depend upon the hardness alone because the structure and the grain size play an important part. It frequently happens that a heat treated piece machines more freely than an untreated forging of the same hardness.

S. A. E. 3145 and 3150 Steels—These steels are not mentioned in the present Society of Automotive Engineers' Handbook, but these numbers are applicable because the steels have the composition of the 3100 series and their average carbon contents are 0.45 and 0.50 per cent respectively. These steels have a greater hardening power than those of the 3100 series having lower carbon contents. S. A. E. 3145 is recommended by Wickenden for large drive shafts and S. A. E. 3150 for oil-hardened gears, the former because the effect of the heat treatment will penetrate deeper into the shaft and the latter because a gear can be produced having a high hardness combined with great strength and toughness.

The heat treatments in the main are the same as for the other steels of this series making allowance for the difference in the critical points. A normalizing treatment is recommended. When difficulty in machining is experienced, it is recommended that the normalizing be followed by reheating to 1250-1350 degrees Fahr. and after due soaking at this temperature be allowed to cool in the furnace. This treatment may be described as tempering rather than annealing. Its object is to produce a fine spheroidized structure. The final quenching and tempering treatments are performed after machining.

S. A. E. 3200 Series Steels—There is a complete series of steels with 0.90-1.25 per cent chromium and 1.50-2.00 per cent nickel with carbon contents up to 0.50 per cent which is known as the S. A. E. 3200 series. These steels are also known as medium chromium-nickel steels. The properties and uses are much the same as the corresponding steels of the S. A. E. 3100 series. As might be expected from the higher content of the alloying elements these steels have greater strength under the same treatments than do the lower chromium-nickel steels. On the other hand they are more expensive. Besides being used for automotive parts the medium carbon steels of this series are used for drop forging

die blocks. Such die blocks are usually quenched all over in oil. If the tempering temperature is high enough, it may be possible to machine the die after heat treating. As was mentioned in the case of the low chromium-nickel steels, the lower carbon steels are usually used for the parts requiring the maximum toughness, while the highest carbon steels in the series are used for such parts as oil-hardened gears which have considerable hardness as well as strength and toughness. In the main, these steels are similar to the S. A. E. 3100 series except that their properties are somewhat superior. On the other hand they do not usually machine as readily as the lower alloy steels. When the carbon content is 0.40 per cent or more it may be necessary to spheroidize after normalizing in order to produce a machinable structure. This treatment has already been described.

S. A. E. 3300 and 3400 Series Steels—These steels may be classified as the high chromium-nickel steels. They are usually employed only for parts requiring exceptional physical properties or very deep penetration of hardness. The general classes of use are the same as those already described. Wickenden recommends the S. A. E. 3400 series as being easier to produce, forge and machine than the S. A. E. 3300 and 3200 series.

With these high chromium-nickel steels it is possible to produce higher values for physical properties than can be produced with the use of most of the other alloy steels. For example the International Nickel Company gives the following values for the physical properties of S. A. E. 3450 steel after oil quenching from above the upper critical point and tempering at 400 degrees Fahr.:

Tensile Strength	274500 lbs. per sq. in. average
Elastic Limit	252500 lbs. per sq. in. average
Elongation	10.5 per cent average
Reduction of Area	38.8 per cent average
Brinell Hardness	500

This particular steel is recommended especially for gears which must withstand heavy tooth pressures on account of the great hardness and strength which can be produced without sacrifice of the necessary toughness. Some authorities prefer the use of a steel such as S. A. E. 2512 for such gears. When that type of steel is used, the gears must, obviously, be case hardened. The question of the relative merits of case hardening versus heat treating of a high alloy medium carbon steel may be discussed later.

High Nickel-Chromium Alloys

Before leaving the subject of nickel-chromium alloys, it may be well to mention some of the high chromium-nickel alloys which are used on account of their resistance to heat, oxidation and corrosion. These vary quite widely in composition for nickel-chromium steels containing about 20 per cent or less of the alloying elements, up to alloys which are composed almost entirely of nickel and chromium with little or no iron or other elements present. The latter, of course, cannot be classified as steels, as they do not have iron as their base.

These alloys are of particular interest to steel treaters as they enter into the equipment of the modern heat treating plants. Among the uses of these alloys may be mentioned:—carburizing pots and boxes, pyrometer protecting tubes, hearth plates, rollers, rails, and other furnace parts subjected to high temperatures, electrical resistors or heating elements for electric furnaces, and thermocouple wires. The use of these alloys as electrical heating elements is not confined to the heat treating room. They may be found in many types of electrical heating apparatus.

The ability of the nickel-chromium alloys to withstand the combined action of heat and air does not seem remarkable when it is considered that both nickel and chromium as pure metals are quite resistant to such actions and the addition of either one of them to steel improves its properties in that respect. Pure nickel crucibles, combustion tubes and boats, have long been used in the chemical laboratory to withstand temperatures up to 1800 degrees Fahr. in the presence of oxygen. The use of nickel plate to withstand the action of weather is well known. In a recent discussion the use of chromium plate was also mentioned.

The uses mentioned for the nickel-chromium alloys depend upon the wonderful ability of these alloys to withstand long exposures to air and furnace gases at temperatures up to 1800-2000 degrees Fahr. without oxidation, scaling, or other harmful effects. It should also be mentioned that some of these alloys are reported as being more or less resistant to the action of some of the common acids and other chemicals.

The accompanying table shows the composition and some of the properties of a number of the commercial nickel-chromium alloys. The wide variation in composition will be readily noted.

Nickel-Chromium Resistant Alloys

Manufacturer	Alloy No.	CHEMICAL COMPOSITION					Form of Material	Melting Point Approx. Deg. F.	Max. Temp. for Continuous Operation	RESISTANCE TO				
		Carbon	Silicon	Manganese	Nickel	Chromium	Iron				Sulphuric Acid		Sea Water and Sea Air	Moist Atmosphere
										Very Dilute	Moderate Dilution	Concentrated		
A	1	From 0.15 to 0.70	From 1.25 to 3.25		From 22.0 to 36.0	From 5.5 to 27.5	From 65.0 to 45.0	Wrought	2700	1800			a	Excellent
	2							Wrought	2700	2100	b	e	d	
	3							Wrought	2700					
B	4	0.80			35.0	15.0	Balance	Cast	2700	2100				
B	5	1.50			6.0	18.0	Balance	Cast	2700					
C	6				61.0	18.5	Balance	Cast	2625	2350				
D	7	0.50	1.50	0.50	25.0	15.0	57.5	Cast	2370					
E	8				60.0	12.0	Balance	Wrought	2300	1830				
E	9				66.0	16.0	Balance	Casting	2400	2010			Excellent	Excellent
E	10				85.0	15.0		Wrought		2100				
E	11				80.0	20.0		Wrought		1110				
E	12				80.0	2.0	Balance	Wrought	2575	2010	Excellent	Excellent	Good	Excellent
					80.0	20.0		Wrought	2570	2010	Excellent	Excellent	Good	Excellent
F	13				85.0	15.0		Wrought	2550	2010	Excellent	Excellent	Good	Excellent
F	14				60.0	16.0	Balance	Wrought	2700	1290	Good	Good	Fair	Fair
F	15				30.0	5.0	Balance	Wrought	2600	2100		f		
F	16				80.0	20.0		Wrought	2550	2010		g		
G	17				85.0	15.0		Wrought	2600	2010				
G	18				Balance	11.0	25.0	Wrought						
G	19				26.0	8.0	66.0	Cast	2600	1110	Good	Good	Excellent	Excellent
G	20				25.0	20.0	55.0							
G	21							Wrought						
H	22	0.45	1.0	0.75	20.0	7.5	Balance							

The data in this table are taken from data sheets accompanying "Combating Corrosion in Industry," *Chemical and Metallurgical Engineering*. Only those alloys containing both chromium and nickel are given. The original tables give various other alloys, also data concerning other properties.

[a] Loss in milligrams per sq. in., 4 days at 175° F.
5 per cent Sulphuric Acid 6.3
10 per cent Sulphuric Acid 1.6
20 per cent Oleum 1.2
30 per cent Oleum 1.2

[b] Loss in milligrams per sq. in., 14 hrs. at 194° F.
100 per cent Sulphuric Acid 0.9
66° Be Sulphuric Acid 0.3
60° Be Sulphuric Acid 19.9
50° Be Sulphuric Acid 85.7

[c] Loss in mill. per sq. in., 10% Sulphuric Acid 194° F. 7 hrs. = 8.4
[d] Loss in mill. per sq. in., 20% Sulphuric Acid 194° F. 7 hrs. = 12.4
[e] Loss in mill. per sq. in., in sea water 8 days at room temp. = .03
[f] Penetration in in. per month room temperature .0004 to .0005
[g] Penetration in in. per month room temperature .0009 to .0010

While prices cannot be mentioned in this discussion, it may be said that the price varies with the amount of the alloying elements contained as both nickel and chromium are more expensive than steel. On the other hand the resistance properties also vary considerably with the composition. The cost to the consumer obviously depends upon the initial cost and the life. It is impossible to give any general conclusion as to which is the best composition as the life depends upon the use to which the alloys are put.

Both castings and forgings are made from the nickel-chromium alloys. Welding by means of the oxyacetylene torch or the electric arc is quite possible although it may be more difficult than the welding of iron, steel and some other metals.

The use of nickel-chromium alloys for carburizing pots and boxes is quite common. While the initial cost of such boxes is much higher than that of cast iron or steel boxes, the long life of alloy boxes often makes them cheaper per hour of service at the carburizing temperature. Some of the manufacturers of alloy pots will guarantee a life of more than 2000 hours at a temperature of 1700 degrees Fahr. A much longer life is frequently reported. It has also been said that in some of the modern continuous carburizing furnaces in which the pots are allowed to cool slowly in the furnace, the life of the alloy pots is very long.

The revolving retorts of the gas carburizing furnaces are also made of nickel-chromium alloys. These retorts must have considerable strength at high temperatures as well as resistance to heat and oxidation. The writer has seen service records on some of these retorts which show a life of more than 7000 hours. On the other hand, some retorts have gone out of service comparatively early on account of casting defects.

The resistance to heat and oxidation is not the only advantage of the alloy carburizing pots. When properly designed such pots or boxes will withstand repeated heating to the carburizing temperature without warping, cracking, growing or undergoing any appreciable change of shape. For this reason it is possible for the boxes and lids to fit tightly during their entire life. This is a decided advantage as it makes for better carburizing. Also as the pots do not change in shape, the mechanical handling is much simplified.

The electric furnace builders have reported that heating ele-

ments of resistant alloys have been in service for several years without being replaced. The use of these heat resisting alloys as heating elements is not limited to the large electric furnaces. Smaller wires of the same type of material are used for many electric heating appliances.

References

American Society for Steel Treating Handbook.
Data Sheets and Reports of the International Nickel Co.
Society of Automotive Engineers Handbook.
Bullens:—"Steel and Its Heat Treatment."
Camp and Francis:—"Making, Shaping and Treating of Steel."
Combating Corrosion in Industry: *Chemical and Metallurgical Engineering* and data sheets.

HIGH SPEED STEELS

(Continued from Page 729)

In the work which Bain and I did on high speed steel, for example, we found only one carbide, or only one other crystalline material, with the X-ray after heating to 1110 degrees Fahr. (600 degrees Cent), and that was definitely as Bain said, a material which, when isolated by chemical solution, showed a high percentage of carbon. So I think it is safe to say, that there are not many of these compounds present in high speed steel at the higher temperatures, and that the main compound which accounts for the greater part of the high temperature hardness, is a complex carbide containing tungsten.

Reviews of Recent Patents

By NELSON LITTELL, Patent Attorney
475 Fifth Ave., New York City—Member of A. S. S. T.

1,621,523, Titanium Alloy, Alvah W. Clement, of East Cleveland, Ohio, Assignor, by Mesne Assignments, to Ludlum Steel Co., of Watervliet, New York, a Corporation of New Jersey.

This patent describes an alloy particularly adapted for resisting the action of corroding gases at high temperatures and, therefore, is useful in making up dies for die castings, carburizing, annealing boxes, and similar purposes. The effective alloy of this type consists of 10 per cent chromium, 6 per cent titanium, 10 per cent aluminum, and the balance iron with possibly some silicon in amounts of less than 1 per cent.

1,621,886, Alloy Steel, William Herbert Keen, of Albany, New York.

The alloy described in this patent is an improvement in high speed steels, designed to improve the cutting qualities and durability of high speed steel tools by increasing both the carbon and vanadium content beyond the proportions usually considered feasible in steels of this character. The new steel may contain

	Per Cent
Carbon	0.75 to 1.20
Vanadium	1 to 5
Tungsten	15 to 19
Chromium	2,50 to 5
Cobalt	0 to 3

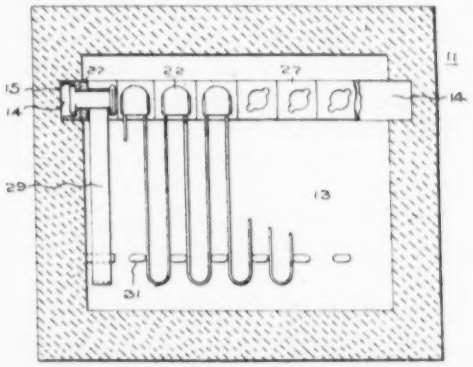
This tool is forgeable and air-hardening and can still retain its cutting edge when working at red heat.

1,622,078, Composition for Treating Metal Baths and Method of Making the Same, Frederick M. Becket, of Douglaston, New York, Assignor to Electro Metallurgical Company, of New York, N. Y., a Corporation of West Virginia.

This patent describes a composition for use in making additions of an alloying element to a molten bath or in removing deleterious ingredients contained therein. The composition is designed to overcome loss due to oxidization when attempts are made to introduce an alloying or scavenging material of a low density into the metal bath. This improvement is brought about by adding and by combining lead with the material to be added to the bath and introducing the compounded lead and addition agent in the form of briquets or the like. The additional weight of the lead causes the briquets to sink rapidly into the bath without any material surface oxidization and in the case of scavengers the compounding with the lead moderates the intensity of the scavenging action. The action of the lead also promotes fluidity in the slag.

1,622,621, **Electric Furnace, James C. Woodson, of East Pittsburgh, Pennsylvania, Assignor to Westinghouse Electric & Manufacturing Co., a Corporation of Pennsylvania.**

This patent shows an electric furnace with removable resistor grids supported on hanger blocks 22 of insulating material provided at the rear part thereof with lugs adapted to extend into the slotted openings in the



channel members 15 supported in the walls of the furnace and to permit the hanger blocks to be turned slightly to lock the hanger blocks by means of said lugs in said openings or to permit removal of the hanger blocks. The lugs are to be opposite the slotted portions of the openings and drawn outward permitting the grids to be removed.

1,623,948, **Nickel Alloy, Truman S. Fuller of Schenectady, New York, Assignor to General Electric Company, a Corporation of New York.**

This patent describes an improvement in nickel-copper alloys designed to increase the tensile strength and elastic property of the alloy. The patent states that alloys such as Monel metal comprising two-third parts of nickel, and one-third parts of copper are desirable for their corrosion resistant properties but have only a moderate use because of the low tensile strength and the elastic limit. The tensile strength and elastic limit may be raised materially by the addition of aluminum and carbon to the nickel-copper alloy. An alloy containing 60 to 75 parts of nickel, 25 to 40 parts of copper, about 2.5 parts of aluminum and 0.16 to 0.17 parts of copper when forged below 1830 degrees Fahr. and suitably heat treated has a tensile strength varying from about 123,000 to 144,000 pounds per square inch and an elastic limit of about 85,000 to 103,000 pounds per square inch, and a reduction of area at breaking varying from 51 to 58 per cent.

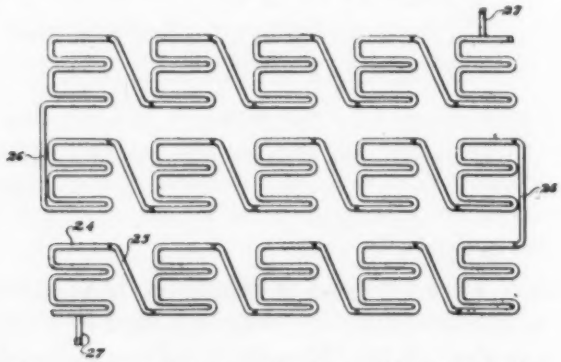
1,624,075, **Steel of High-Temperature Stability, William Robert Shimer, of Bethlehem, Pennsylvania, Assignor to Bethlehem Steel Company.**

The object of this invention is to improve the physical properties of steel at high temperatures particularly to provide for the addition to the steel of an agent which preserves at high temperatures the physical characters of a steel having high tensile properties at normal temperatures. The addition of chromium to steel to increase its tensile strength in the untreated state is well-known. Chromium steels, however, cannot be sub-

mitted to high temperatures without losing their high tensile strength. This inventor has discovered that a small addition of tungsten in conjunction with chromium modifies the strength at high temperatures on the steel permitting it to retain its high tensile strength at the higher temperatures. A preferred composition for the new alloy is 0.43 per cent chromium, 1.30 per cent tungsten, 0.64 per cent manganese, 0.33 per cent carbon, and the balance iron. The addition of tungsten up to 2 per cent in the case of low carbon steels containing no chromium also renders the steel more stable at high temperatures.

1,624,217, Electric Furnace, Frank T. Cope of Salem, Ohio, Assignor to the Electric Furnace Company, of Salem, Ohio, a Corporation of Ohio.

This patent describes an electrical furnace of the type in which the resistors or grids are mounted in the roof and floor of the furnace. The major improvement comprises producing the resistors, as castings, of sufficient rigidity and mechanical strength at high temperatures to resist distor-



tion so as to make it possible to eliminate the insulating separators now used to space the resistors of drawn or rolled wire or ribbon. The drawings show a plan view of a floor grid having the sinuous units 24 connected by angular strands 25 and end strands 26, and having the usual terminals 27. This grid is preferably cast of heat resisting alloy metal and may be of any structural cross-sectional shape.

1,612,765, Furnace, Frank M. Lee, Chicago, Illinois, Assignor to Sullivan Machinery Company.

This patent covers a heating furnace consisting of an elongated combustion chamber having a horizontal plane bottom and an enlarged portion for initially receiving the combustible mixture and an aligned elongated communicating restricted heating portion. A narrow elongated passage connects with the restricted portion of the combustion chamber throughout the length of the latter and through which the burned mixture from the chamber passes. A burner discharges longitudinally of the chamber, the plane bottom and restricted portion graduating the heat to a maximum temperature at the end of the chamber remote from the burner discharge.

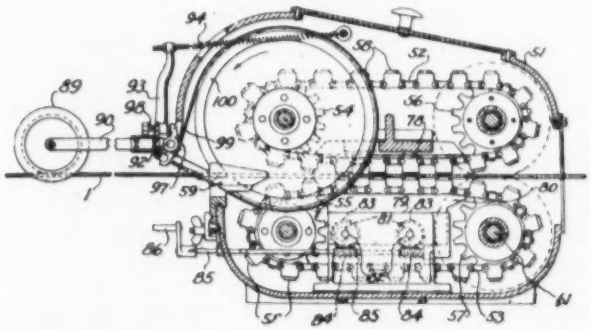
1,610,809, Electric Furnace, Delbert F. Newman, Schenectady, New York, Assignor to General Electric Company, New York.

The furnace charge is supported within a container in which an atmosphere

of selected gas is maintained. A portion of the charge can be withdrawn from the underside of the closure, which is provided for the container, and a new portion inserted without exposing the remaining portion of the charge within the container to the effects of air.

1,624,668, **Apparatus for Heat-Treating Metals, Frederic S. Kochendorfer, of River Forest, Illinois, Assignor to Western Electric Company, Incorporated of New York, N. Y., a Corporation of New York.**

This patent describes a heat treating apparatus for heating and cooling the metal in a continuous manner within closely regulated temperature limits. The apparatus shows means for feeding metallic strands without subjecting the strands to any substantial tension and also shows means for introducing a non-oxidizing and non-reducing gas into the chambers and



for measuring the electrical characteristics of the strand continuous with the treatment thereof. The drawings comprise 13 figures showing a means for continuously passing the strands of metal through the treating apparatus which consists essentially of supply reels from which strands of metal are unwound and fed forward by caterpillar feeding mechanism such as illustrated in the accompanying figure through a series of heating furnaces and of heating and cooling tubes to a second caterpillar feeding device and thence on to the storage reels.

1,617,859, **Method of Annealing Copper Articles, Sterner St. P. Meek, of Philadelphia, Pennsylvania.**

This patent describes a method of annealing copper articles, which comprises packing the articles in powdered metallic copper or copper dust, heating to the desired temperature, excluding atmosphere and then cooling.

1,612,571, **Smelting Furnace, Henry L. Charles, Kent, Washington.**

In this smelting furnace, the hearth is provided with a convexly shaped integral bank extending along each side and end wall and terminating adjacent to the roof. An integral mound extends longitudinally between the banks at the side walls, one end of the mound being contiguous with the bank at the firing end, and the other end semiconical in form and spaced away from the bank at the flue end.

1,612,690, **Oil-Gas Apparatus, Jean Zwicky, Farnham, Royal, England.**

This patent covers a partial-combustion oil-gas-producing apparatus com-

prising a liquid fuel jet having an air passage surrounding the jet with a chamber for the liquid fuel surrounding the air passage and connected to the jet. An annular air conduit encircles the fuel chamber and leads to the air passage.

1,619,462, Manufacture of Alloy Steel and Iron, Byramji D. Saklatwalla, of Crafton, Pa.

This patent describes the method of making alloy steel without introducing additional carbon in the metal of the bath when the alloying material is added. In the usual practice it is customary to add the alloying material as a reduced metallic element in the form of a ferro-alloy which usually contains a considerable amount of carbon, unless a very high grade alloy is used, which carbon when added to the bath increased the carbon content of the bath beyond the point desired. The present patent describes a process of adding the alloying material without increasing the carbon content which comprises maintaining a bath of steel having a metal layer and a slag layer and incorporating successive charges of a reducing agent into the metal layer the oxidation products of which renders the slag more fluid and incorporating into the slag layer successive charges of an unreduced compound of the alloying metal and maintaining the bath at a temperature to cause a reaction which will directly reduce the alloying metal into the bath.

1,621,472, Mold for Casting Metals, William H. Smith of Detroit, Michigan, assignor to Ford Motor Company of Highland Park, Michigan, a Corporation of Delaware.

This patent describes a metal mold of the "permanent type" for the casting of iron and other materials therein. The metal of the mold should have less than 1 per cent of foreign substances therein. A suitable composition is "armco" iron.

THE ENGINEERING INDEX

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Arrangements have been made with The American Society of Mechanical Engineers whereby the American Society for Steel Treating will be furnished each month with a specially prepared section of The Engineering Index. It is to include items descriptive of articles appearing in the current issues of the world's engineering and scientific press of particular interest to members of the American Society for Steel Treating. These items will be selected from the copy prepared for the annual volume of the Index published by the A. S. M. E.

In the preparation of the Index by the staff of the A. S. M. E. some 1,200 domestic and foreign technical publications received by the Engineering Societies Library (New York) are regularly searched for articles giving the results of the world's most recent engineering and scientific research, thought, and experience. From this wealth of material the A. S. S. T. will be supplied with a selective index to those articles which deal particularly with steel treating and related subjects.

Photostatic copies (white printing on a black background) of any of the articles listed may be secured through the A. S. S. T. The price of each print, up to 11 by 14 inches in size, is 25 cents. Remittances should accompany orders. A separate print is required for each page of the larger periodicals, but whenever possible two pages will be photographed together on the same print. When ordering prints, identify the article by quoting from the Index item: (1) Title of article; (2) name of periodical in which it appeared; (3) volume, number, and date of publication of periodical; and (4) page numbers.

AIR HEATERS

COMBINED GAS AND AIR ADMISSION. Combined Gas and Air Admission for Air Heaters (Kombinerte Gas-und Lufteinführung für Winderhitzer), J. Stoecker. Stahl u. Eisen, vol. 47, no. 12, Mar. 24, 1927, pp. 493-494, 3 figs. Describes system of combined gas and air admission, advantages of which are low initial cost, safety against explosion; good mixture of gas and air, takes up little space, and is very durable; system has been used successfully for number of years on air heaters for three blast furnaces in Bochumer Verein, Bochum, Germany.

ALLOY STEELS

LOCOMOTIVES. Making Locomotives Stronger, W. A. Newman and C. F. Pascoe. Iron Age, vol. 119, no. 10, Mar. 10, 1927, pp. 701-704, 10 figs. Canadian railroad secures better quality and service from vanadium and nickel cast-steel frames; manufacture, structure and heat treatment.

TITANIUM, DETERMINATION IN. Determination of Small Quantities of Titanium in Highly Alloyed Steels (Die Bestimmung kleiner Mengen Titan in hochlegierten Stählen), K. Roesch and W. Werz. Chemiker-Zeitung, vol. 51, no. 16, Feb. 26, 1927, pp. 149-150. Points out that for determination of small quantities of titanium, titration is best method; separation of titanium from iron is effected according to Rothe ether method.

ALLOYS

INTERNAL FRICTION. Methods of Measurement and the Internal Friction of Bismuth-Tin Alloys and the alloy Cu_3Sn (Weiteres über die Messmethode und über die innere Reibung der Bi-Sn-Legierungen und der Legierung Cu_3Sn), F. Sauerwald and K. Töpler. Zeit. für anorganische u. allgemeine Chemie, vol. 157, no. 1-3, Oct. 26, 1926, pp. 117-137, 7 figs. Internal friction is determined by viscosity measurements; describes improved apparatus in which fused metal is drawn through capillary, time for definite volume to pass at given temperature being noted. See brief translated abstract in Brass World, vol. 23, no. 3, Mar., 1927, p. 89.

ALUMINUM

DISCOVERY. The "Discovery" of Aluminum, M. Tosterud and J. D. Edwards. Am. Electrochem. Soc.—Advance Papers, April 28-30, 1927, no. 10, 4 pp. Oersted's classic experiments are described; 2 years later Wöhler modified Oersted's experiment, using potassium metal in place of potassium amalgam to reduce $AlCl_3$; recently Fogh verified Oersted's results; dilute potassium amalgam works best; Tosterud and Edwards again repeated experiments of Oersted and found no difficulty in obtaining aluminum metal analyzing 99 per cent Al.

ELECTRIC WIRING. The Durability of Aluminum Overhead-Line Wires (Ueber die Haltbarkeit von Freileitungsdrähten aus Aluminium), O. Hähnel. Elektrizitätswirtschaft, vol. 26, no. 428, Mar. 1, 1927, pp. 101-103. Experience of German Post Office authorities with aluminum wire for overhead lines; in order to possess required mechanical strength, aluminum must be alloyed with small admixtures of other metals, although these admixtures deprive aluminum of a great deal of its resistivity against effect of atmosphere; aluminum wire experimentally used by German Post Office did not prove satisfactory; hard copper wire is much more durable.

FINISHING PROCESS. Finishing Process for Aluminum and Its Alloys (Sur un nouveau procédé d'ornementation de l'aluminium et de ses alliages), L. Guillet. Académie des Sciences—Comptes Rendus, vol. 184, no. 3, Jan. 17, 1927, pp. 134-136. Pacz has shown that if aluminum or alloy rich in aluminum is tempered in solution of sodium fluosilicate, nickel sulphate and

potassium nitrate in 4 liters of water at 70 to 80 degrees, colored lines appear which blacken quickly in bath; design produced may be varied according to relative motion with respect to liquid accorded to metal; deposits appear to start from impure spots on surface, and their extent is strictly limited by electrolytic protection; corrosion readily starts along lines of pattern, with formation of alumina.

GASKET MATERIAL. Aluminum as a Gasket Material, J. E. Housley. Power, vol. 65, no. 10, Mar. 8, 1927, p. 374. Aluminum gaskets have been used for many purposes about oil refineries for number of years owing to resistance to corrosion and oxidation by oil compounds at high temperatures and pressures.

ALUMINUM ALLOYS

ALUMINUM-SILICON. Aluminum-Silicon Alloys, H. Ehlermann. Eng. Process, vol. 18, no. 3, Mar., 1927, pp. 75-76, 3 figs. Discusses wide use of these alloys owing to their excellent casting properties which increase tensile strength by means of refining process; employment for pressure die castings and for structural work.

CABLES AND WIRES. Aluminum Alloy, Almelec of High Rupturing Strength (Un nouvel alliage léger de haute resistance mécanique pour fils et câbles conducteurs: l'almelec), E. Dusaugey. Revue Générale de l'Electricité, vol. 21, no. 8, Feb. 19, 1927, pp. 303-305. New alloy has recently been found, containing 98.5 per cent Al and 1.2 per cent Mg and Si; wires drawn from these alloy ingots receive special hardening and heat treatment process, which gives them up to 50 per cent greater strength than copper; accompanying tabulation gives characteristics of alloy in comparison with copper and pure aluminum; "almelec" makes cables which are stronger and lighter than electrically equivalent steel-aluminum cable.

CASTING. Light Aluminum Casting Alloys. Brass World, vol. 23, no. 3, Mar., 1927, p. 76. List of aluminum casting alloys presenting fairly complete arrangement of all alloys ordinarily used in aluminum foundries, and furnishing mixtures which may prove more suitable than those at present used by some establishments.

CORROSION. The Corrosion of Aluminum Alloys, H. Sutton. Metallurgist (Supp. to Engineer), Mar. 25, 1927, pp. 36-37. Resistance of aluminum to corrosion by ordinary industrial waters and sea water is in striking contrast to its position in table of electrode potentials of metals and its chemical reactivity which indicate that aluminum should be very rapidly attacked; actually, attack is rather slow and frequently local in character; resistance is due to presence of protective film of aluminum oxide or hydroxide, and to capacity of metal to combine chemically to form this protective film on freshly cut or scraped surfaces; intercrystalline corrosion; aluminum and its alloys corrode under marine conditions most severely when corrosion product is allowed to accumulate on surface; anodic oxidation process constitutes marked advance in protection of aluminum and its alloys; other methods.

FOUNDRY SCRAP. Secondary Metal Versus Scrap in Aluminum-Alloy Foundry Practice, R. J. Anderson. Am. Metal Market, vol. 34, no. 53, Mar. 17, 1927, pp. 3-6, 4 figs. Discusses case for and against general use of scrap aluminum and aluminum alloys in foundry, and states present case as regards use of secondary aluminum and aluminum alloys; use of primary metal in castings production.

WROUGHT. Wrought Light Alloys, W. Rosenhain. Metallurgist (Supp. to Engineer), Mar. 25, 1927, pp. 39-40, 1 fig. Review of present position in regard to light alloys available for structural purposes in rolled and forged condition; question arises as to whether duralumin is still best available structural light alloy; review of other alloys which makes it clear that, however useful and meritorious duralumin has proved, it has by no means attained maximum that can be anticipated from aluminum alloys.

AUTOGENOUS WELDING

LABORATORY RESEARCH. Role of the Laboratory in Autogenous Welding (Le role du laboratoire en soudure autogène), A. Roux. Technique Moderne, vol. 19, no. 4, Feb. 15, 1927, pp. 105-110, 18 figs. Points out that physical and chemical phenomena which accompany operations of welding are extremely complex; describes laboratory which is well designed and organized, especially for work of research and testing; chemical and physico-chemical tests; X-ray examination; mechanical tests of welds; bending and impact tests.

REPAIR WORK. Autogenous Welding Methods for Repair Work, C. C. Custer. Nat. Engr., vol. 31, no. 4, Apr., 1927, pp. 169-170, 5 figs. Points out that welding is not applicable in all cases; repairing material to be welded.

BEARING METALS

PRODUCTION AND REFINING. The Production and Refining of White Bearing Metals. Engineering, vol. 123, no. 3191, Mar. 11, 1927, pp. 282-284, 2 figs. Influence of copper in low-tin bearing metals; in bearing metals high in tin, lead is considered impurity which may be present in comparatively small amounts only; melting-pot plant for cleaning and alloying of white bearing metals; suitable scrap metals; how copper is removed from scrap; methods of refining.

BERYLLIUM

PROPERTIES. The Light Metal Glucinium (Le Glucinium, métal léger). Nature (Paris), no. 2757, Mar. 15, 1927, p. 251, 1 fig. Metal has crystalline aspects; it is hard, of steel-gay color and susceptible of high polish; it is resistant to atmospheric agents and to hot water, covering itself with protective film; it is only slightly soluble in alkaline and acids; density is 1.84; melting point in neighborhood of 1380 deg.; Brinell hardness, 140.

BLAST FURNACES

LEAD. The Treatment of Lead Residues, E. R. Thews. Metal Industry (Lond.), vol. 29, nos. 15 and 20, Oct. 8, and Nov. 12, 1926, pp. 333-335 and 457-458. Operation of lead blast furnace.

OXIDATION PHENOMENA. More Iron From Wide Hearths. Iron Age, vol. 119,

no. 12, Mar. 24, 1927, pp. 856-857. Influence of oxidation phenomena on blast furnace process; reactions before tuyeres; possible improvements indicatel.

REACTIONS. Some Blast Furnace Reactions Analyzed, G. Tobin. Blast Furnace & Steel Plant, vol. 15, no. 3, Mar., 1927, pp. 114-116. "Solution loss" and air volume calculated; by equations basis for gas formation is rendered apparent; detailed method of determining air volumes is given.

SINTERING PLANTS. Sinter for Blast Furnace Burdens, E. J. Tournier. Iron Age, vol. 119, no. 11, Mar. 17, 1927, pp. 775-778, 5 figs. Improvements in treating ferrous materials by Greenawalt intermittent process; oil or gas ignition.

BOILER PLATE

EMBRITTLEMENT. The Embrittlement of Boiler Plate, F. G. Straub. Forging-Stamping-Heat Treating, vol. 13, no. 3, Mar., 1927, pp. 89-95, 10 figs. Reasons for formation of cracks are clearly outlined; micrographic studies aid in determining character of cracks; causes and prevention of embrittlement.

BOILERMAKING

BOILER-DRUM PRODUCTION. Production of Boiler Drums for High-Pressure Boilers (Die Herstellung von Kesseltrommeln für Hochdruckdampfkessel), B. Schapira. Wärme u. Kälte-Technik, vol. 28, no. 24, Dec. 1, 1926, pp. 287-289, 4 figs. Describes process of welding drums with water gas, employed at August Thyssen Works; seamless forging of drums at works of F. Krupp, and at Reisholz rolling mills at Düsseldorf.

BOILER-DRUM PRODUCTION. The Manufacture of High-Pressure Boiler Drums. Engineer, vol. 143, no. 3712, Mar. 4, 1927, pp. 246-248, 11 figs. As alternative to forging, firm of Thyssen & Co. has developed process of making seamless drums by welding, which, it is claimed, produces equivalent results at about two-thirds the expense; details of process of manufacture and testing employed by this firm in production of drums for very high pressures.

BOLTS

CASE-HARDENING. The Development of Soft Spots in Case-Hardened Bolts (Ueber die Entstehung von weichen Stellen an Bolzen, die im Einsatz gehärtet sind), H. Graefe. Werkstattstechnik, vol. 21, no. 4, Feb. 15, 1927, pp. 89-90, 3 figs. By means of metallographic investigations of structure, it is shown that soft spots are due to effect of annealing in polishing; results of tests carried out in experimental station of Siemen-Schuchert Works.

BRASS

ELECTRIC MELTING. Twenty-Five Years of Nonferrous Electrothermics; Fifteen Years of Electric Brass Melting, H. W. Gillett. Am. Electrochem. Soc.—Advance Papers, Apr. 28-30, 1927, no. 24, 20 pp. Electrothermics finds some application in annealing and heat treatment of nonferrous alloys, but its chief application is in melting; proportion of electrically melted metal is much greater in brass industry than in steel industry; electric brass melting is now practically standardized upon 3 specific types of furnaces.

TIN AND SOLDER PENETRATION. Penetration of Brass by Tin and Solder, H. J. Miller. Brass World, vol. 23, no. 3, Mar., 1927, pp. 73-75, 1 fig. Cracking of stressed brass articles by intercrystalline penetration of molten solder; experiments indicate high stresses required for penetration.

BRONZES

FOUNDING. Bronze Founding, J. Dewrance. Engineer, vol. 143, no. 3713, Mar. 11, 1927, pp. 274-276, 20 figs. Bronze when cooling from fluid state crystallizes, size of crystals depending on time occupied in process; period of crystallization is governed by temperature of metal poured into mold, mass or thickness of casting, and cooling properties of mold. Presidential address before Institute of Metals.

NICKEL-ALUMINUM. Nickel Aluminum Bronzes, F. D. Gleason. Metal Industry (N. Y.), vol. 25, no. 4, Apr., 1927, p. 149. Alloy is superior to phosphor, manganese and gun bronzes and special steels as set forth by U. S. Navy on specifications for inspection of materials; sand-cast materials; various bronzes and their properties.

CALORIMETERS

METAL. An Improved Metal Calorimeter for Determination of Specific Heat of Metals, Oxides and Slag (Ein verbessertes Metallkalorimeter zur Bestimmung der spezifischen Wärme von Metallen, Oxyden und Schlacken), W. Grosse and W. Dinkler. Stahl u. Eisen, vol. 47, no. 11, Mar. 17, 1927, pp. 448-453, 6 figs. Instrument for determination of specific heat in temperature ranges of 100 to 1600 deg.; calibration of calorimeter; results of tests; device is combination of Oberhoffer vacuum process with Nernst-Lindermann calorimeter.

CAST IRON

ABRASIVE RESISTANCE. The Resistance to Wear of Cast Iron in the Case of Sliding Friction. Foundry Trade Jl., vol. 35, no. 549, Feb. 24, 1927, pp. 173-174. Results of experiments carried out by O. H. Lehmann. Translated from Giesserei-Zeitung, nos. 21, 22, 23, Nov. 1, 15 and Dec. 1, 1926. See reference to original article in Eng. Index, A. S. S. T. Trans., Feb., 1927, p. 307.

AUTOMOTIVE INDUSTRY. Cast Iron in Its Relation to the Automotive Industry, E. J. Lowry. Soc. Automotive Engrs.—Jl., vol. 20, no. 2, Feb., 1927, pp. 277-290, 13 figs. Data presented tend to prove that true hardness is not measurable by any known test; further, that present hardness test is not function of machinability or of wear and that combined carbon bears little relation to any of these three factors; governing features seem to be forms of carbon, structure of iron and quality of materials used; method used in determining quality of material is that of dilation; this test showed that expansion of casting is related to expansion of material used; it is felt that continual experimentation with this method will enlighten many unanswerable questions which arise concerning cast iron at present.

ELECTRIC MELTING. Gray Cast Iron from the Point of View of the Electric Furnace, G. K. Elliot. West. Machy. World, vol. 18, no. 3, Mar., 1927, pp. 122-123 and

136. Correct understanding of chemistry of both acid and basic-hearth electric furnaces should lie at the foundation of every decision for or against adoption of electric furnace in competition with cupola; points out difference in behavior of basic and acid furnaces toward molten cast iron.

GRATE BARS. Cast Iron Alloy Developed for Grate Bars. Ry. Mech. Engr., vol. 101, no. 3, Mar., 1927, pp. 181-182, 2 figs. Placed on market under trade name of Nogroth metal, an air-furnace product; its structure resembles that of white cast iron and has many properties of cast steel; it retains its original strength at high temperatures, nor will it grow when exposed to temperatures below melting point.

HEAT TREATMENT. The Heat Treatment and Growth of Cast Iron, J. W. Donaldson. Foundry Trade Jl., vol. 35, nos. 548 and 549, Feb. 17 and 24, 1927, pp. 143-146 and 167-169 and (discussion) 169-171, 13 figs. Account of series of heat treatment tests carried out on iron cast by filter process; results show that carbide decomposition accompanied by graphite deposition takes place in gray cast iron when subjected to low-temperature heat treatment; stability of carbide is influenced by silicon content, and also by initial combined carbon content; also by temperature and duration of heat treatment.

IMPROVEMENT. The Refinement of Cast Iron by Mechanical Process. Fuels & Furnaces, vol. 5, no. 4, Apr., 1927, pp. 501-502. Mechanical process described by M. Irresberger before Society of German Foundrymen, consists in subjecting liquid cast iron to violent shaking or vibrating motion, which activates and accelerates such reactions as are known to occur in bath; liquid iron is degasified and deoxidized, and all components of cast iron are thoroughly mixed, thus neutralizing harmful effects of sulphur; graphite is dissolved and more finely-grained structure is obtained.

OVERHEATING. Effect of Extended Overheating on Structure and Properties of Cast Iron (Einwirkung einer weitgehenden Ueberhitzung auf Gefüge und Eigenschaften von Gusseisen), F. Meyer. Stahl u. Eisen, vol. 47, no. 8, Feb. 24, 1927, pp. 294-297, 10 figs. Effect of graphitization through overheating in molten state on degree of fineness; account and results of tests, and conclusions therefrom; it is shown that in all cases, with increasing overheating of pig and cast iron, increased graphitization occurred.

PHOHPHORUS, EFFECT OF. Influence of the Phosphorus Content of Cast Iron on the Resistance to Compression and Tensile Stresses, W. E. Dennison. Foundry Trade Jl., vol. 35, no. 552, Mar. 17, 1927, pp. 229-230. Results of experiment carried out to determine deterioration of gray cast iron when phosphorus content is raised from 0.8 to 1.2 per cent.

SHEAR TESTS. Ring Shear Tests for Cast Iron, M. Rudeloff. Metallurgist (Supp. to Engineer), Mar. 25, 1927, p. 47, 1 fig. Results of experimental investigation; although ring shear test as originally proposed can be performed more rapidly and with smaller specimens than more usual mechanical tests, it nevertheless requires specimens about 20 mm. in length and 25 mm. in diameter from which test piece of special shape has to be machined; investigation has been extended to include simplified punch test, in which all that was required was thin disk of material about 2mm. in thickness; expressions obtained may be used with good accuracy to deduce, from punch tests, other mechanical properties for cast irons having punch shear strengths ranging from 18 to 32 kg. per sq. mm. Translated from Giesserei, vol. 13, Aug. 14 and 21, 1926.

STRUCTURAL CONSTITUENTS. The Calculation of Structural Constituents from Cast-Iron Analysis (Die Berechnung der Gefügebestandteile aus der Gusseisenanalyse), H. Pinsl. Giesserei, vol. 14, no. 11, Mar. 12, 1927, pp. 161-165, 3 figs. Attempt is made to calculate structural constituents as they occur in gray cast iron.

SULPHUR IN. Limitation of Sulphur Effect for Ordinary Cast Iron. Soc. Mech. Engrs. (Japan)—Jl., vol. 30, no. 118, Feb., 1927, pp. 47-74, 15 figs. Determination of permissible amount of sulphur in ordinary iron castings for machine construction, mostly by means of mechanical and microscopic tests; for soft iron (over 2 per cent of Si) permissible amount of sulphur is up to 0.12 per cent; for hard iron (about 1.5 per cent of Si) up to 0.10 per cent, and for very hard iron (under 1.0 per cent of Si) up to 0.08 per cent. (In Japanese.)

CASTING

BRONZES. Remember Those Who Go Down to the Sea in Ships. Foundry, vol. 55, no. 7, Apr. 1, 1927, pp. 259-263, 20 figs. Methods employed in casting monument of bronze fisherman set up in public square in Gloucester, Mass.; pattern was divided into eight sections; how sectional parts were handled; pouring castings.

CHROMIUM

PROPERTIES AND USES. Chromium—Its Properties and Uses, C. G. Fink. Soc. Automotive Engrs.—Jl., vol. 20, no. 1, Jan., 1927, pp. 157-160. More important characteristics are extreme hardness, exceptional resistance to high temperatures and corrosion and high reflectivity; because it imparts toughness to other metals, about 45 per cent of production is used in alloys; alloy of chromium with nickel, iron and manganese, called nichrome, is used as electric resistor in enameling and heat treating; Goldschmidt process of producing chromium metal consists in mixing green chromium oxide with aluminum which acts as reducing agent; stainless steel is made from 13 per cent chromium and iron.

CHROMIUM-NICKEL STEEL

CARSIL PROCESS. The Carsil Process. Foundry Trade Jl., vol. 35, no. 552, Mar. 17, 1927, p. 235. Manufacture of high-grade nickel-chromium steel from Java iron sand at works of C. G. Carlisle & Co., Sheffield.

MOLYBDENUM IN. Molybdenum in Nickel-Chromium Medium Carbon Steels. Chem. Age (Met. Sect.), vol. 16, no. 401, Mar. 5, 1927, pp. 17-19. Investigation of influence of molybdenum, carried out by Research Department, Woolwich; thermal critical ranges; microstructure and heat treatment; mechanical properties.

COPPER

ARSENIC IN. The Effect of Arsenic on Copper, D. Hanson and C. B. Marryat. Engineering, vol. 123, no. 3191, Mar. 11, 1927, pp. 303-306, 22 figs. Experiments show that solubility of arsenic at all temperatures is greater than 7 per cent; further annealing at 650 deg. cent. for 96 hours failed to dissolve constituent completely; solubility of 7.25 per cent of arsenic in solid copper was indicated, and this solubility changes very little with fall of temperature. Abstract of paper read before Institute of Metals.

ARSENIC PLUS OXYGEN IN. The Effect of Arsenic Plus Oxygen on Copper, D. Hanson and C. B. Marryat. Engineering, vol. 123, no. 3191, Mar. 11, 1927, pp. 306-308, 5 figs. Results of tensile tests; annealed rod, tested at ordinary temperature and at 250 deg. cent.; cold-rolled rod; experiments show that small amounts of oxygen have distinct embrittling effect which may be corrected by arsenic; Brinell hardness; fatigue range; electrical conductivity; macrostructure of cast billets and microstructure; effect of annealing in hydrogen. Abstract of paper read before Institute of Metals.

BRIGHT DIPPING. Bright Dipping of Metal, A. K. Graham. Am. Electrochem. Soc.—Advance Papers, Apr. 28-30, 1927, no. 23, 9 pp. Discusses application of acid dips to copper and brass, and acid mixtures frequently recommended; study of chemistry of dips containing sulphuric, hydrochloric and nitric acids has been made; it was found that rate of solution of copper varies directly with nitric acid concentration, while rate of solution of zinc varies directly with concentration of hydrochloric acid; activity of dip depends upon proportion of sulphuric acid to water, and upon temperature.

ELECTROLYTIC REFINING. The Specific Resistivity of Copper Refining Electrolytes and Method of Calculation, S. Skowronski and E. A. Reinoso. Am. Electrochem. Soc. —Advance Papers, Apr. 28-30, 1927, no. 11, 18 pp. Specific resistivity of sulphuric acid at different concentrations and at various temperatures was determined; effect of addition of various amounts of copper, nickel, arsenic and iron on specific resistivity of sulphuric acid was studied and from results obtained it was found that within limits of composition of electrolyte used in copper refining, effect of copper, nickel, arsenic and iron on increasing specific resistivity of electrolyte is directly proportional to amounts added.

PROPERTIES. Copper as Engineering Material (Kupfer als Werkstoff), P. Melchior. V. D. I. Zeit., vol. 71, no. 12, Mar. 19, 1927, pp. 373-379, 8 figs. Production of crude copper; structure and diagram of state; chemical and physical properties; influence of annealing and of impurities; defects; uses.

WIRE, RECRYSTALLIZATION. Recrystallization of Copper Wire (Ueber die Rekristallisation von Kupferdraht), E. Schmid and G. Wassermann. Zeit. für Physik, vol. 40, no. 6, Dec. 14, 1926, pp. 451-455, 6 figs. Copper wire hard drawn shows no change in X-ray photographs until heated at about 100 deg., when recrystallization begins, simple thread-like structure being formed with direction parallel to axis of wire; such wire possesses no elastic recoil.

COPPER-IRON-SILICON. Reaction and Equilibrium in Copper-Iron-Silicon System with Special Regard to Copper Matte (Reaktionen und Gleichgewichte im System Cu-Fe-Si mit besonderer Berücksichtigung des Kupfersteins), O. Reuleaux. Metall u. Erz., vol. 24, nos. 5 and 6, Mar. 1 and 2, 1927, pp. 99-111 and 129-134, 24 figs. Binary systems, Cu-Fe, Cu-Si and Fe-Si; crystallized state; solidification phenomena. Bibliography.

CORSON. Copper Hardened by New Method, M. G. Corson. Brass World, vol. 23, no. 3, Mar., 1927, pp. 77-79, 3 figs. What Corson alloys are; stronger cable wire possible; many uses suggested; silicon-aluminum and silver-silicon alloys.

CORROSION

LIGHT ALLOYS. Corrosion Tests with Hydrogen Peroxide—Sodium Chloride Solutions (Korrosionsversuche mit alkalisch-reagierender Wasserstoffsuperoxyd-Kochsalzlösung auf verschiedene Leichtmetall-Legierungen), E. Rackwitz and E. K. O. Schmidt. Korrosion u. Metallschutz, vol. 3, no. 1, Jan., 1927, pp. 5-10, 7 figs. Results of tests show that Mylius oxidic salt test as well as author's modified method do not as yet meet requirements for rapid corrosion testing of light alloys; first method could perhaps be improved by certain changes recommended by author.

RESEARCH. Present Status of Corrosion Research (Heutiger Stand der Korrosionsforschung), J. Hausen. Stahl u. Eisen, vol. 47, no. 7, Feb. 17, 1927, pp. 272-274, 1 fig. Investigation of different phenomena, showing that no general rule can be given for prevention of metal corrosion; only detailed knowledge of behavior of every single metal can serve as basis for its proper use under given conditions.

INTERNAL DIFFUSION. Temperature of Commencement of Internal Diffusion in Crystals (Die Temperatur des Beginns innerer Diffusion in Kristallen), G. Tammann. Zeit. für anorganische u. allgemeine Chemie, vol. 157, no. 4, Nov. 16, 1926, pp. 321-325. Sintering together of powders on heating is most probably due to molecules commencing to change their places; the more complicated the molecule is, the nearer does this temperature approach melting point.

PLASTIC DEFORMATION. Increase of Tensile Strength of Single Crystals by Plastic Deformation (Ueber die Schubverfestigung von Einkristallen bei plastischer Deformation), E. Schmid. Zeit. für Physik, vol. 40, no. 1-2, Nov. 29, 1926, pp. 54-74, 8 figs. Increase of strength of main faces of slip in zinc crystals caused by plastic extension has been investigated; after passing elastic limit, strength increases linearly with extension. See brief translated abstract in Brit. Chem. Abstracts, Feb., 1927, p. 99.

CUPOLAS

FLUORSPAR ADDITIONS. Fluorspar in Cupola Practice. Iron Age, vol. 119, no. 14, Apr. 7, 1927, pp. 997-998. German tests indicate that it does not act as desulphurizer where lime used is pure; further research needed. Abstract translated from Stahl u. Eisen, Jan. 27, 1927, p. 128.

STEEL-SCRAP MELTING. The Melting of Steel Scrap in Cupolas (A propos de la

Fusion des Riblons d'acier au Cubilot), M. Guedras. Fonderie Moderne, vol. 21, Feb., 1927, pp. 35-36. Remarks based on article in Oct. 1926 issue of same pournal.

CUTTING TOOLS

ROUGH TURNING. Rough Turning with Particular Reference to the Steel Cut, H. J. French and T. G. Digges. Mech. Eng., vol. 49, no. 4, Apr., 1927, pp. 339-352 and (discussion) 352-354, 17 figs. Tests extend to current commercial high-speed tool and structural alloy steels portions of Taylor's original investigations in rough turning carbon steels made primarily to show effect upon tool performance of variation in chemical composition and mechanical properties of steel cut. See reference to complete article in A. S. M. E. advance paper in Eng. Index Section of A. S. S. T. Trans., Jan., 1927, p. 137.

DIES

CLASSIFICATION. Classifying Dies in Stamping Shop, P. J. Edmonds. Forging-Stamping-Heat Treating, vol. 13, no. 3, Mar., 1927, pp. 87-88. Extended outline of general division of dies with further subdivision in detail; terms employed are defined.

ELECTRIC FURNACES

BRASS-MELTING. Electricity in the Foundry, C. R. Burch and N. R. Davis. Metropolitan Vickers Gaz., vol. 10, no. 167, Feb., 1927, p. 47. Describes electrical performance of furnace for melting brass and copper; it was designed to take type A50 salamander crucible.

ELECTROTHERMIC PROCESSES. The Use of Electric Furnaces at Niagara Falls 1902 to 1926, F. A. J. FitzGerald. Am. Electrochem. Soc.—Advance Paper, no. 2, for mtg. Apr. 28-30, 1927, pp. 47-50, 3 figs. Presents figures showing remarkable growth of furnaces used in certain electrothermic processes in Niagara Falls during 25 years; electric-furnace processes are divided into three groups, namely: abrasives, carbon products and ferroalloys and calcium carbide.

HEAT TREATING. Tubular Electric Furnace for Heat Treating, H. O. Swoboda. Forging-Stamping-Heat Treating, vol. 13, no. 3, Mar., 1927, pp. 96-97, 1 fig. Embodies radical departure from prevalent pusher type.

HIGH-FREQUENCY. The Study of High-Frequency Induction Furnaces (Zur Kenntnis des Hochfrequenz-Induktionsofens), F. Wever and W. Fischer. Mitteilungen aus dem Kaiser-Wilhelm-Institut für Eisenforschung, vol. 8, nos. 69 and 70, parts 10 and 11, 1926, pp. 149-170 and 171-179, 47 figs. Paper no. 69: Theory and design of ironless induction furnaces; current and field distribution in interior of solid cylinder; air transformer; general aspects of construction and examples of actual designs; possibilities of development. Paper no. 70: Metallurgy of ironless induction furnaces; decarburization and deoxidation; production of carbon-poor and carbon-rich steels in ironless furnace; refining tests with pure nickel; possibilities of development.

HIGH-TEMPERATURE. Tungsten Spiral Furnace for Very High Temperatures (Wolframspiralöfen für sehr hohe Temperaturen), W. Fehse. Zeit. für Technische Physik, no. 3, 1927, pp. 119-122, 5 figs. Results of tests for purpose of improving heaters.

INDUSTRIAL APPLICATIONS. Electricity in Industrial Heating. Iron Age, vol. 119, no. 13, Mar. 31, 1927, pp. 918-919. Heat treating departments need improving; forging furnaces have faults; advantages in melting gray iron and brass electrically.

NONFERROUS. Electric Furnaces in Nonferrous Metallurgy, D. F. Campbell. Metal Industry (Lond.), vol. 30, nos. 11 and 12, Mar. 18 and 25, 1927, pp. 287-289 and 309-311, 4 figs. Deals in practical way with principal electric furnaces used for melting alloys of copper, nickel, zinc and silver, with special reference to induction furnaces of normal and high frequency. Paper read before Institute of Metals. See also Engineering, vol. 123, no. 3192, Mar. 18, 1927, pp. 334-336, 4 figs.; and Foundry Trade Jl. vol. 35, no. 551, Mar. 10, 1927, pp. 209-212, 4 figs.; and editorial comment, p. 207.

NONFERROUS. Electric Furnaces for Nonferrous Metals. Metallurgist (Supp. to Engineer), Mar. 25, 1927, pp. 35-36. Review of papers and discussion presented before Institute of Metals; great difficulty which stands in way of wider application of high-frequency principle to industrial melting is first cost and upkeep of plant required for generation of high-frequency current; progress made in America in melting of larger masses of metal; question of electric annealing furnaces.

PRACTICE. German Engineers Discuss Electrical Furnace Practice. Foundry Trade Jl., vol. 35, no. 549, Feb. 24, 1927, pp. 175-177. Review of following papers; Chemical Reactions in Basic Electric Furnace by F. von Kerpely. Economy of Electric Furnace in Foundry, by E. Kothny; Coreless Type of Induction Furnace, by W. Fischer; Charges for Electrical Energy, by W. Schaefer.

ELECTRIC WELDING, ARC

BUILDING CONSTRUCTION. Wide New Field Opens for Welding, G. E. Hagemann. Mfg. Industries, vol. 13, no. 3, Mar. 1927, pp. 215-216, 4 figs. Westinghouse puts up large five-story building by arc-welding process; advantages offered through medium of welded construction.

BUTT-WELDING MACHINE. Lincoln Carbon-Arc Butt-Welding Machine. Am. Mach., vol. 66, no. 14, Apr. 7, 1927, p. 599, 1 fig. Recent development of Lincoln Electric Co., Cleveland; it is claimed that this welder can be applied to certain classes of butt-welding problems that cannot be handled on resistance welder.

STEEL PIPES. Arc Welded Steel Pipe for High Pressure Mains for New York City, W. Schenstrom. Am. Welding Soc., Jl., vol. 6. no. 2, Feb. 1927, pp. 30-36. First section of large-size diameter, electrically welded water pipe to be installed in large city in United States; it is 48-in. I. D. pipe made of ½ in. flange steel, welded inside and out; results of tests.

STRUCTURAL STEEL. Arc Welded Structural Steel Buildings, C. G. Watson. Iron & Steel Engr., vol. 4, no. 3, Mar. 1927, pp. 151-155, 9 figs. and (discussion) 155-169, 34 figs. Describes erection of 100-per cent welded steel building by Youngstown Welding Company; it is of standard mill type; not a rivet or bolt was used in fabrication or erection; even derrick used in erection was welded.

STUDY OF. Study of Electric Autogenous Welding (Etude sur la soudure electrique autogene), H. Dustin. Revue Universelle des Mines, vol. 12, no. 5, Dec. 1, 1926, pp. 177-206, 22 figs. Results of research including study of structure and mechanical properties of welding metal and action of welding on basic metal.

ELECTRODES

GRAPHITE AND CARBON. Graphite and Amorphous Carbon Electrodes in the Iron and Steel Industry (Graphitelektroden und amorphe Kohleelectroden in der Eisen- und Stahlindustrie), W. Bliemeister. Stahl u. Eisen, vol. 47, no. 11, Mar. 17, 1927, pp. 457-458. Presents table of relative values of carbon and graphite electrodes; great differences between the two are shown in heat conductivity; strength is approximately the same for both; advantages and disadvantages of both types.

FORGING

MACHINE. Progress in Machine Forging, C. D. Harmon. West. Machy. World, vol. 17, no. 12, Dec., 1926, pp. 534-535, 10 figs. and vol. 18, no. 2, Feb., 1927, pp. 64-65 and 91. Comparison of relative merit of machine forging and drop forging operations.

MACHINES. Forging Machines (Les machines à forger), L. Gendron. Pratique des Industries Mécaniques, vol. 9, no. 12, Mar., 1927, pp. 485-494, 20 figs. Points out defects of older types; safety devices; improvements in machine; machines for low-temperature forging; recent innovations.

METHODS. Forging Two Unusual Parts. Am. Mach., vol. 66, no. 14, Apr. 7, 1927, pp. 581-582, 4 figs. Replacing difficult casting with forging; forming in one operation; hot drawing very heavy shell; parts are being made at East Pittsburgh plant of Westinghouse Electric & Manufacturing Co.

STEEL. Some Notes on the Forging of Steel, O. W. Ellis. Metallurgist (Supp. to Engineer), Mar. 25, 1927, pp. 46-47. Deals first with factors affected by forging machines, namely, energy and velocity of blow; and then with factors determined by material being forged, namely, volume of metal being forged and its carbon content.

FOUNDRIES

PRACTICE. Old-Fashioned Methods and Improvements in Foundry Practice (Rückständigkeiten und Fortschritte im Giessereiwesen), T. Ehrhardt. Giesserei, vol. 14, no. 10, Mar. 5, 1927, pp. 145-151, 6 figs. Layout and equipment of large modern foundry; improvements of unsatisfactory equipment; new materials; foundry for special-machine works; general arrangement of auxiliary plants; iron recovery from scrap; improving quality of gray iron castings.

FUEL ECONOMY

IRON AND STEEL INDUSTRY. Fuel Economy in the Iron and Steel Industry, H. A. Brassert. Fuels and Furnaces, vol. 5, no. 4, Apr., 1927, pp. 467-471. Discussion of more economical use of fuels; proper application of fuel; cleaning of gas; gas enrichment; use of preheated air; effect of gas constituents on heat transmission.

METALLURGICAL INDUSTRY. Fuel Economy and the Metallurgical Industry, E. W. Smith. Gas World, vol. 86, no. 2223, Mar. 12, 1927, pp. 240-242. Notes on waste-heat boilers; use of coke oven and blast furnace gas; insulation by aluminum paint.

FURNACES, ANNEALING

GAS-FIRED. Modern Gas-Fired Annealing Furnaces, T. Teisen. Foundry Trade Jl., vol. 35, no. 552, Mar. 17, 1927, pp. 238-239. Suiting producer to fuel; regenerative and recuperative furnaces; malleable annealing furnaces.

FURNACES, GAS

FORGING AND HARDENING. Thermal and Operating Economy in Forging and Hardening (Wärmewirtschaft und Betriebswirtschaft in Schmieden und Härtereien), K. Kassler. Sparwirtschaft, no. 3, Mar., 1927, pp. 123-125. Points out advantages of gas firing: in spite of 50 per cent additional cost, gas furnaces effect saving of 5.65 per cent; for mass production of parts economically constructed continuous furnaces can be employed.

RADIANT-HEAT. Gas-Fired Furnaces for High Temperature Work. Gas Age—Rec., vol. 59, no. 14, Apr. 2, 1927, pp. 488-489, 2 figs. Describes system of furnace construction which was first given publicity in Gesolei in Duesseldorf, Germany, where several examples of furnaces were exhibited in 1926; these are known as Krupp radiating furnaces containing radiant body, and are made and sold by Mindoga Industrieöfen und Gaserzeuger Gesellschaft, m. b. H., Essen, Germany.

REGENERATIVE. The Siemens Smelting and Copper Refining Furnaces and Their Advantages over Semi-Gas Furnaces (Die Siemens-Schmelz- und Kupferraffinieröfen und ihre Vorzüge gegenüber Halbgasöfen), A. Sprenger. Metall u. Erz, vol. 24, no. 6, Mar. 2, 1927, pp. 121-128, 16 figs. Based on practical results, advantages of Siemens regenerative gas furnaces for refining of copper and metal-containing ores are set forth; comparison with semi-gas furnaces; main points in favor of these gas furnaces are low coal consumption, easy control and high efficiency and durability; discusses possibility of increasing efficiency by use of mechanically charged tilting furnace and of gas-heated auxiliary equipment.

FURNACES, GAS

ROTATING. A Rotating Gas Furnace, J. Williams. Am. Mach., vol. 66, no. 14, Apr. 7, 1927, pp. 592-593, 1 fig. Furnace in forge shop of Morse Twist Drill & Machine Co. is used to heat drill blanks that are to be upset on one end to form tapered square shanks on "bit-stock" drills.

FURNACES, HEAT-TREATING

DESIGN. Practical Industrial Furnace Design, M. H. Mawhinney. Forging-Stamping-Heat Treating, vol. 13, no. 3, Mar., 1927, pp. 83-86 and 88, 2 figs. Construction of furnaces for heat treating plays important part in success of operation; heating processes are explained by examples.

TYPES. Heat Treatment and Metallography of Steel, H. C. Knerr. Forging-Stamping-Heat Treating, vol. 13, no. 3, Mar.,

1927, pp. 103-107 and 110, 3 figs. Furnace materials; essentials for good heating; types of furnaces; electric furnaces.

FURNACES, HEATING

REVERSIBLE REGENERATION. The Influence of Flue Gas Temperature upon Fuel Consumption. Metal Industry (Lond.), vol. 30, no. 9, Mar. 4, 1927, pp. 238-239, 2 figs. Attempt has been made during past few years by Davis Furnace Co., Luton, to reduce heat losses on gas-fired furnaces by adoption of system of reversible regeneration applied to air supply; design of furnace in which this principle is adopted is known as Revergen; advantages of system; typical results available upon annealing of cast iron and steel in furnaces of Revergen type.

FURNACES, INDUSTRIAL

AIR DISTRIBUTION. Air Supply for Industrial Furnace (Windlieferung für Industrieöfen), Maercks. Fördertechnik u. Frachtverkehr, vol. 20, no. 6, Mar. 18, 1927, pp. 121-124, 5 figs. Air pipe; individual or group distribution; regulation of blast capacity; control of air supply; parallel connection of ventilators.

FURNACES, METALLURGICAL

EFFICIENCY. Definition of Thermal Efficiency for Metallurgical Furnaces (Wirkungsgrad metallurgischer Ofen), P. Rosin. Metall u. Erz, vol. 24, no. 4, Feb. 2, 1927, pp. 73-81, 17 figs. Relationship between calorific value and flue-gas volumes of all fuels and calorific value-volume diagrams; J-t diagram for flue gas and its practical application.

SMALL. Most Important Types of Small Furnaces (Die wichtigsten kleinofenarten), E. Hasselblatt. Werkstattstechnik, vol. 21, no. 4, Feb. 15, 1927, pp. 98-103, 15 figs. Brief review of development; most important difference between small and large metallurgical furnaces; most important types, dealing especially with hardening, annealing, forging and melting furnaces.

HARDNESS

ROCKWELL AND BRINELL NUMBERS. Relationships Between the Rockwell and Brinell Numbers, S. N. Petrenko. U. S. Bur. Standards—Technologic paper no. 334, Jan. 10, 1927, pp. 195-222, 7 figs. Comparative Rockwell and Brinell tests were made on great variety of ferrous and nonferrous metals; experimental values of Brinell and Rockwell numbers were inserted into given equations and constants determined; these theoretical equations with experimentally determined constants may be used to estimate, within error of plus or minus 10 per cent, Brinell number from Rockwell number; similar equations were obtained for tensile strengths of steels, which may, within an error of plus or minus 15 per cent be estimated from Rockwell number.

ROCKWELL TEST. Rockwell Hardness-Test Approved. Soc. Automotive Engrs.—Jl., vol. 20, no. 1, Jan., 1927, pp. 13-14. Present S. A. E. specifications cover only Brinell and Shore hardness tests; Rockwell tester measures hardness by determining depth of penetration of steel ball or diamond cone (Brale) in material being tested under conditions of load application.

IMPACT TESTING

NOTCHED-BAR. The Influence of Size of Specimen and Temperature of Stress-Deformation Curve in Notched-Bar Tests (Ueber den Einfluss der Probenbreite und der Temperatur auf den Kraftverlauf beim Kerbschlagversuch), F. Körber and H. A. v. Storp. Mitteilungen aus dem Kaiser-Wilhelm-Institut für Eisenforschung, vol. 8, part 8, no. 67, 1926, pp. 127-134, 11 figs. Describes modified method for determination of stress-deformation curves; determination of maximum load and deflection from these curves for carbon steel at different temperatures and with different cross sections of test piece; influence of temperature on working constants and working speeds.

IRON

DESULPHURIZING. Desulphurization of Iron by Fluorides. Metallurgist (Supp. to Engineer), Mar. 25, 1927, pp. 37-38, 1 fig. Review of experiments by Wilke-Dörfurt and Buchholz, described by them in Stahl u. Eisen, Dec. 23, 1926; they begin with idea that desulphurization should be more rapidly brought about by action of fluorides, which are more readily decomposed than is relatively stable calcium fluoride; results are given in diagram, and it will be seen that effect is greatest with fluorides which are most readily decomposed; desulphurizing action of fluorides was even more marked when fluorides acted on iron alone than when reaction took place under or in presence of large amount of slag. See reference to original article in Eng. Index Sec. of A. S. S. T. Trans., Mar., 1927, p. 481.

MAGNETIZATION LOSSES. Apparatus for the Determination of Losses and Magnetizability of Iron, T. B. Vlassov. Electritchestvo, no. 2, Feb., 1927, pp. 67-74, 11 figs. Describes two apparatus, constructed by author, one, for determination of losses in iron, convenient for big factory tests, second, for testing magnetizability of iron, constructed on principles of permeameter of Hopkinson. (In Russian.)

RUSTLESS. Rustless Iron by a New Process. Iron Age, vol. 119, no. 14, Apr. 7, 1927, pp. 990-992, 3 figs. Novel method of melting developed in Sweden; electrodes of ferrochromium suggested; chromium losses reduced; carbon increases kept down.

SOLUBILITY AND HARDNESS. Changes in Solubility and Hardness of Metallic Materials Due to Torsion and Bending (Ueber einige durch Tordieren und Biegen hervorgerufene Veränderungen der Löslichkeit und Härte metallischer Materialien), G. Carre. Korrosion u. Metallschutz, vol. 3, no. 1, Jan., 1927, pp. 1-5, 4 figs. Results of torsion tests; change in solubility and in ball impression; tests of cold-bent iron bars; results show that in twisted as well as bent iron, in contrast to cold-drawn iron, blowholes occur which by heating at 900 degrees Cent. do not disappear, and therefore, are injurious to material, both mechanically and chemically.

STAINLESS. Characteristics of Enduro Stainless Iron. Am. Mach., vol. 66, no. 10, Mar. 10, 1927, p. 427. Physical properties of Enduro bars. Reference-book sheet.

TRANSFORMER. Influence of Treatment of Transformer Iron on its Watt Losses (Ueber den Einfluss der Behandlung des

Transformatoreneisens auf seine Wattverluste), G. Eichenberg and W. Oertel. Stahl u. Eisen, vol. 47, no. 7, Feb. 17, 1927, pp. 262-271, 21 figs. Review of existing literature; process in production of transformer plate with 4 per cent silicon; conditions affecting quality of plate; results of rolling tests to determine influences of degree of deformation, total heat duration, etc.; determination of relations between grain number, grain form and watt-loss coefficient.

IRON ALLOYS

IRON-CARBON. The Diagram of State of Iron-Carbon Alloys and its Applications (Das Zustandschaubild der Eisenkohlenstoff-Legierungen und seine Anwendung), Hanemann. V. D. I. Zeit., vol. 71, no. 8, Feb. 19, 1927, pp. 245-253, 31 figs. Describes cementite, graphite and martensite systems; results of metallographic investigations.

IRON CASTINGS

CHILLED. Hematite as a Base for Chilled Castings? Foundry Trade Jl., vol. 35, no. 551, Mar. 10, 1927, p. 221. It can be safely stated that in majority of cases, cold-blast iron is more suitable as base than hematite iron.

OXIDIZING. Oxidizing Gray Iron Castings, W. S. Barrows. Can. Foundryman, vol. 18, no. 3, Mar., 1927, p. 16. Castings must be polished and plated before they can be properly oxidized.

PRODUCTION. Iron Castings and Their Production, H. Allen. Commonwealth Engr., vol. 14, no. 6, Jan. 1, 1927, pp. 225-228. From engineer's point of view, the higher the tensile strength of cast iron, the lighter castings can be made; sulphur below 0.1 per cent does not much affect castings, but when present in higher proportions tends to keep carbon in combination, thus resulting in hard castings which are difficult to machine; design of cupolas; reactions in melting zone of blast furnace.

IRON, PIG

MOLTEN. The Internal Friction of Molten Pig Iron (Ueber die innere Reibung von flüssigem Roheisen), H. Thielmann and A. Wimmer. Stahl u. Eisen, vol. 47, no. 10, Mar. 10, 1927, pp. 389-399, 14 figs. Results of tests with open and with closed viscosimeter; molecular theory; by means of comparative measurements, friction coefficients of certain iron-carbon alloys were determined, and at the same time relation of internal friction to temperature was investigated; results are shown graphically in form of isotherms.

SPECIFICATIONS. Specifications for Pig Iron. Metallurgist (Supp. to Engineer), Mar. 25, 1927, pp. 38-39. It is well known that chemical analysis alone does not completely define pig iron; purchase of pig iron to analysis must be carefully differentiated from purchase to specification; while arguments against rigid purchasing specifications of national character for pig iron are perfectly sound, they in no way affect desirability of specifications for cast iron.

MAGNESIUM ALLOYS

MELTING AND CASTING. Melting and Casting of Magnesium Alloys (Das Schmelzen und Vergiessen von Magnesiumlegierungen), H. Reininger. Giesserei, vol. 14, nos. 2 and 12, Jan. 8, and Mar. 19, 1927, pp. 17-21, and 182-190, 53 figs. Review of developments in use of magnesium alloys, with special reference to electron metals; melting and casting methods. Mar. 19: Melting equipment; economy of different melting processes and plants; casting of magnesium alloys; methods and equipment.

MALLEABLE CASTINGS

BLACK-HEART. The Cupola and Black-Heart Malleable (Le cubilot at la malléable à coeur noir; sa pratique), C. Kluystmans. Fonderie Moderne, vol. 21, Feb., 1927, pp. 45-51. Deals with cupola process in melting pig which is to be annealed for obtaining black-heart malleable; deals with air volume, pressure and calculation of charge; temperature and composition; points out that black-heart malleable can be produced at lower price in cupola than in reverberatory furnace.

FITTINGS. Malleable Iron Pipe Fittings, H. R. Hiscott. Machy. Market, no. 1375, Mar. 11, 1927, pp. 19-20, 2 figs. Discusses fittings which are screwed British standard gas or pipe thread, and are for use with gas, water or steam barrel; describes typical shop in which such fittings are made.

METALLOGRAPHY

IRON AND STEEL. The Properties of Iron and Steel as Affected by Composition and Heat Treatment, H. M. Boylston. Fuels and Furnaces, vol. 5, no. 4, Apr., 1927, pp. 433-440, 12 figs. Discussion on metallography of iron and steel; effect of carbon content on physical properties; relation of carbon content to uses of steel.

METALS

ELECTROLYTIC REFINING. Electrolytic Refining During the Past Twenty-Five Years, L. Addicks. Am. Electrochem. Soc.—Advance Papers, Apr. 28-30, 1927, no. 13, 3 pp. While much progress has been made in practice, considerably greater progress has been made in theory of electrolytic refining; 25 years ago copper was practically the only metal electrolytically refined on large commercial scale; today lead, zinc, gold and silver are refined by very similar processes; electrolytic treatment of liquors obtained by leaching of copper ores is now obtaining wide application and will continue to expand rapidly.

FATIGUE. Some Modern Views on the Fatigue of Metals, H. J. Gough. Structural Engr., vol. 5, no. 3, Mar., 1927, pp. 70-81 and (discussion) 81-83, 7 figs. Attrition theory of Ewing and Humfrey; Bauschinger's theory of natural range of elastic limits; stress-strain relations under safe range and under unsafe range of stress; quasi-elasticity or elastic hysteresis; effect of slip caused by stress concentrations; effects of static and repeated stresses on microstructure of metals; theory of strain hardening governing behavior of metals under static and repeated stresses. Bibliography.

FUSION HEAT. Latent Heat of Fusion (Zur Theorie der Schmelzwärmen), N. von Raschevsky. Zeit. für Physik, vol. 40, no. 3-4, Dec. 6, 1926, pp. 214-219, 2 figs. Proceeding from assumption that liquid state is reached when distance between atom-

becomes so great that elastic modulus becomes zero, second differential of lattice energy in respect to lattice constant becoming zero, value of latent heat of seven metals is calculated and is in satisfactory agreement with experiment.

GROWTH. Metal Growth Causes Motor Starter to Fail, D. A. Hampson. Machy. (N. Y.), vol. 33, no. 8, Apr., 1927, p. 601. Instance in which motor starter that had stood idle for a time failed to work, as result of "metal growth" in one of parts; it is well known that white metal parts change size under atmospheric conditions, in much the same manner as iron castings, and instance referred to emphasizes this characteristic.

HEAT- AND ACID-RESISTING. Heat and Acid-Resisting Materials (Hitze- und säurebeständige Konstruktionsstoffe), B. Schulz. Brennstoff- u. Wärmewirtschaft, vol. 9, no. 3, Feb. 1, 1927, pp. 62-65, 2 figs. Discusses progress made in solution of this problem before and during World War, and recent post-war investigations and experimental results of Post-War period.

MELTING POINT. Relation Between Temperature of Incipient Increase in Size of Granules and Melting Point of Metals (Ueber die Beziehung der Temperatur des Beginns des Kornwachstums zu der Schmelztemperatur der Metalle), A. A. Botschwar. Zeit. für anorganische u. allgemeine Chemie, vol. 157, no. 4, Nov. 16, 1926, pp. 319-320. For number of metals ratio of temperature at which new granules commence to be formed in metal, worked as far as possible in cold, to melting point tends to increase with decrease in latter temperature, but usually lies between 0.35 and 0.45.

POURING PROCESS. A New Process for Pouring Zinc and Other Metals, W. M. Corse. Am. Zinc Inst.—Bul., vol. 10, no. 2-3, Feb.-Mar., 1927, pp. 23 and 50, 1 fig. One of most successful processes for overcoming difficulties of obtaining clean ingot is that known as Durville process, invented by French engineer; application of process to casting aluminum and its alloys has been very successful; it has been found, however, that other metals such as zinc and copper are greatly improved by use of this process.

RECRYSTALLIZATION. Recrystallization (Ueber Rekristallisation), H. Hanemann. Stahl u. Eisen, vol. 47, no. 12, Mar. 24, 1927, pp. 481-489 and (discussion) 489-491, 14 figs. Review of literature; process and speed of recrystallization; recrystallization diagrams; maximum of recrystallization; theoretical and practical conclusions.

TENSILE TEST. The Tensile Test—Some Less Well-Known Features, G. Sachs. Metallurgist (Supp. to Engineer), Mar. 25, 1927, pp. 44-46, 8 figs. Tensile stress as plotted in conventional load-extension diagram is only nominal, actual or true stress at any point being greater in proportion to decrease in cross-section of test piece; true stress increases from beginning to end of test; maximum load occurs at commencement of local necking of specimen; tests show that for mild steel, copper and aluminum, order of occurrence of phenomena is (1) local necking, (2) maximum load, (3) end of uniform elongation; in practice it is found that effect of local necking extends for some distance, diameter of test piece gradually increasing from point of fracture; nor can end effect be neglected for materials having large extension; accurate determination of uniform elongation can in such cases only be made on long bars. Translated from Berichte der Fachausschüsse des Vereins deutscher Eisenhüttenleute.

MICROSCOPES

METALLIC. A New Metallic Microscope (Ein Neues Metallmikroskop), H. Freund. Giesserei, vol. 14, no. 8, Feb. 19, 1927, pp. 109-110, 1 fig. Describes new metallic microscope which very nearly fulfills requirements demanded of large Chatelier microscope, although this microscope is small apparatus, which for economic reasons are usually preferred in foundries.

NONFERROUS METALS

MICROSCOPIC INVESTIGATIONS. Microscope is Useful Adjunct in Nonferrous Investigations, J. S. G. Primrose. Foundry, vol. 55, no. 6, Mar. 15, 1927, pp. 222-226 and 235, 19 figs. Author points out rules to be observed and admonishes investigator to examine specimens under low power of magnification before etching so that general structure may be seen before resorting to high magnification of special constituents. Abstract of paper presented before Institute British Foundrymen, 1926.

TEMPERATURE DETERMINATION. Judging Temperature by Sight, R. R. Clarke. Can. Foundryman, vol. 18, no. 3, Mar., 1927, pp. 5-6. Appearance of various nonferrous metals at various temperatures is discussed, and practical information given to enable molder to judge actual temperature of metal under widely varying conditions.

OPEN-HEARTH FURNACES

END PORTS. Open-Hearth End Ports, B. Finney. Iron Age, vol. 119, no. 10, Mar. 10, 1927, p. 715, 2 figs. By deviation from standard practice in construction, basic open-hearth steel furnace at Lima, Ohio, plant, has been enabled to produce four 25-ton heats in 24 hr.; this oil-burning furnace has three instead of customary two uptakes.

GAS-PRODUCER ATTACHMENT. High Temperature Obtained in Small Open-Hearth, H. G. Begeman. Foundry, vol. 55, no. 6, Mar. 15, 1927, pp. 209-210, 2 figs. New furnace, developed by P. Ostendorf, differs radically from other types in that gas producers were incorporated into body of furnace itself, one at each side; it works with excellent results in Austrian steel foundry in making various kinds of steel; temperature of gas, which is mixed in gas port with products of combustion formed in gas producer, is increased considerably as it is exposed to high temperature of gasification zone; due to this arrangement, temperatures over 3300 degrees Fahr. may be reached in furnace, which makes it possible to pour steel castings with extremely thin walls.

HEAT DISCHARGE FROM REFRACTORY LININGS. Determination of Heat Discharge Through Refractory Linings of Furnace (Bestimmung des Wärmeabflusses durch das feuerfeste Mauerwerk von Feuerungen), K. Huffelmann. Feuerungstechnik, vol. 15, no. 10, Feb. 15, 1927, pp. 109-113, 7 figs.

Description of experimental furnace and results obtained; deviation of rule-of-thumb formula for determination of external and wall temperature; tests were carried out with blast-furnace gases.

RECUPERATORS. Recuperators for the Open-Hearth, W. H. Fitch. Iron Age, vol. 119, no. 13, Mar. 31, 1927, pp. 920-922, 3 figs. Suggestion for recovering part of waste heat now going up stacks; less fuel and more steel claimed possible; carborundum used in construction.

OXYACETYLENE WELDING

OPEN-HEARTH FURNACES. Cutting and Welding at the Open-Hearth. Blast Furnace & Steel Plant, vol. 15, no. 3, Mar., 1927, pp. 139-141. Use of oxyacetylene blow pipe facilitates tapping, opening "frozen" nozzles and reclaiming scrap; welding equipment makes possible emergency repairs.

PIPE LINES. Pipe Line Welding from the Oxyacetylene Viewpoint, L. Edwards. Am. Welding Soc.—Jl., vol. 6, no. 2, Feb., 1927, pp. 7-24, 18 figs. Advantages; notes on preparation, material, aligning of pipe, methods of laying pipe, testing of line, expansion and contraction, etc.

STEEL PLATE. Oxyacetylene Welding Steel Plate, T. C. Fetherston. Am. Welding Soc.—Jl., vol. 6, no. 2, Feb., 1927, pp. 25-30, 4 figs. Two major factors of importance are fusion and penetration; former indicates thorough mixing of molten metal from sides of joint and from welding rod, and latter indicates careful welding of pieces to be joined entire way to bottom of space between edges.

PISTONS

NICKEL-ALLOY. Nickel Alloy Pistons Improve Performance. Motorship, vol. 12, no. 4, Apr., 1927, pp. 299-300, 3 figs. Engine power can be increased, wear on bearings is reduced, acceleration improved and smoother running obtained with Bu-nite nickel-alloy pistons made by Butler Mfg. Co. of Indianapolis from alloy of nickel, aluminum and copper produced by special process which results in complete union of metals.

RAILS

FRACTURES. Internal Cracking of Rails with Oval Spots (La fissuration interne des rails, avec tache ovale), L. Pichard. Genie Civil, vol. 90, no. 5, Jan. 29, 1927, pp. 118-123, 13 figs. Presents theory of crack formation; homogeneity and heterogeneity in a rail; graphic representation of heterogeneity of a rail section; results of deformation; damages in rails due to rolling process; formation of transversed fissures due to simple heterogeneity of metal, etc.

JOINTS, WELDED. Progress Report Number Four Repeated Impact Tests. Committee on Welded Rail Joints. Am. Bur. Welding, Am. Elec. Ry. Eng. Assn., July., 1926, 35 pp., 20 figs. Deals with cast iron, electric resistance (bar and butt), electric seam, and thermit welded joints; relation of American Electric Railway Engineering Association to committee; repeated impact tests of welded rail joints; explanation of charts showing progressive failure of joints under repeated impact tests; report of subcommittee on inspection of fractured joints; Bureau of Standards metallographic report; discussion in impact testing.

ROLLING MILLS

BAR MILLS. Mill to Roll Bars, Rods and Strip, F. L. Prentiss. Iron Age, vol. 119, no. 11, Mar. 17, 1927, pp. 779-783, 8 figs. Continuous sheet bar mill run by 9000-horsepower motor; flexibility is outstanding feature of new 10-inch merchant mill of Corrigan, McKinney Steel Co., Cleveland.

ELECTRIFICATION. Electrification of Phoenix Mills (Pa.), R. H. Wright. Iron Age, vol. 119, no. 10, Mar. 10, 1927, pp. 710-712, 3 figs. All steam drives and steam lines eliminated; structural-mill capacity increased 25 per cent.

ELECTRIFICATION. Electrification of Steel Mills in Sweden (Variable motordrift inom järnindustrien), F. Freden. Teknish Tidskrift (Elektroteknik), vol. 57, no. 1, Jan. 8, 1927, pp. 1-4. General review of progress of electrification, systems most commonly in use and probable trend of future developments; general data are given for 68 electrified mills; 36 of these have drives with special speed regulation; of these 36 mills 16 utilize induction motors with cascade regulation, 12 D. C. (Ward Leonard) and 8 induction motors with slip regulation (Krämer system).

INDUCTION MOTORS FOR. Speed Regulation, Compounding and Phase Compensation in Three-Phase Induction Motors for Steel-Mill Applications (Varvtalsreglering, kompoundering och faskompensation av trefasasynkronomotorer), L. Dreyfus. Teknisk Tidskrift (Elektroteknik), vol. 57, no. 1, Jan. 8, 1927, pp. 4-9, 15 figs. Author enumerates advantages of motor with speed regulation: (1) speed-torque characteristics can be varied over considerable range without affecting efficiency; (2) no-load current can be selected arbitrarily; if so desired, motor can be used for power-factor correction; (3) load factor can be predetermined for entire speed range; (4) overload capacity can be increased considerably; considers advantages and disadvantages of various systems of steel-mill drives in which induction motors are used.

LAYOUT AND EQUIPMENT. Rolls Wide Range of Products, R. A. Fiske. Iron Age, vol. 119, no. 13, Mar. 31, 1927, pp. 923-926, 6 figs. Mill at Wisconsin Steel Works is equipped to produce universal plates, structural shapes, rounds, squares and flats. See also description in Iron Trade Rev., vol. 80, no. 13, Mar. 31, 1927, pp. 832-834, 4 figs.

MANNESMANN PROCESS. Mannesmann Process (Das Schrägwalzen), F. Kochs. Stahl u. Eisen, vol. 47, no. 11, Mar. 17, 1927, pp. 433-446, 30 figs. Principle of process; stresses in rolled material, mechanical and dynamic phenomena; magnitude, speed and direction of acting forces; twisting of ingot is unavoidable result; influence of mandrel on hole formation; tests, with aid of macroscopic and microscopic examinations, on deformation phenomena. Bibliography.

PLATE MILLS. Thin-Plate Mills (Das Feinblechwalzwerk), W. Krämer. Stahl u. Eisen, vol. 47, nos. 6 and 10, Feb. 10 and Mar. 10, 1927, pp. 209-219 and 399-406, 42 figs. Discusses most important methods of producing thin plate in cold and hot mills; details of such mills, including dry and platinum heating furnace; comparison of German and American rolling methods;

lubrication of bearings; heat treatment of plate; suggestions for improvements in order to obtain greater efficiency.

SEMI-STEEL

PROPERTIES. Semi-Steel, J. E. Hurst. Foundry Trade Jl., vol. 35, nos. 552, and 553, Mar. 17, and 24, 1927, pp. 231-232 and 257-258. Mar. 17: Composition; total carbon contents. Mar. 24: Industrial semi-steel; low total-carbon semi-steels; character of steel additions; soundness and regularity of final castings.

SHEET METAL

THIN PLATE. Production and Use of Thin Plate in United States of America (Herstellung und Verwendung von Feinblechen in den Vereingten Staaten von Nordamerika), H. Klein. Stahl u. Eisen, vol. 47, no. 7, Feb. 17, 1927, pp. 259-262. Comparison of production and consumption data for United States and Germany; American propaganda for use of thin plate; methods and equipment of rolling mills.

SPRINGS

TEMPERING. Tempering and Resetting Springs, C. A. Kyle. Elec. Ry. Jl., vol. 69, no. 13, Mar. 26, 1927, pp. 576-577, 1 fig. Spring repairs are facilitated and cost of tempering and resetting springs reduced by spring-tempering furnace of gas-burning type installed in shops of New York State Railways, Syracuse, N. Y.

STEEL

AIRCRAFT CONSTRUCTION. Commercial Aeronautics and Steel, Wm. P. MacCracken, Jr. Iron Age, vol. 119, no. 14, Apr. 7, 1927, pp. 987-989, 3 figs. Increasing use of ferrous metals in aircraft building; requirements, purposes and limitations.

COLD WORKING. Cold Working and Quench-Hardening. Metallurgist (Supp. to Engineer), Mar. 25, 1927, pp. 47-48. Review of article by Houdremont & Bürklin in Stahl u. Eisen, Jan. 20, 1927, dealing with hardening of carbon steels by quenching and cold working; correspondence of density changes with hardness when produced by cold work and quenching respectively, was not found to be complete; hardness of cold-drawn steel was considerably lower than that hardened by quenching with same change of volume; on internal-stress hypothesis, localization of effects of cold work cannot be readily understood.

CORROSION. Influence of Rust-Film Thickness Upon the Rate of Corrosion of Steels, E. L. Chappell. Indus. & Eng. Chem., vol. 19, no. 4, Apr. 1927, pp. 464-466, 1 fig. In absence of rust films different commercial steels corrode in water or in atmosphere at characteristic rates determined by chemical properties of their surfaces; natural course of atmospheric corrosion does not lead to formation of heavy rust films, so copper-bearings steels have been shown to be superior for this service; underwater corrosion generally leads to formation of heavy rust film, and little difference has been found in corrosion rates of steels under water; there is quantitative decrease in corrosion rate about proportional to increases of rust-film thickness.

DISTORTION IN. Distortion in Heat-treated, Casehardened Carbon Steels, P. J. Haler. Junior Instn. Engrs., vol. 37, Feb. 1927, pp. 216-231, 8 figs. Casehardening process; considerations relating to distortion; pressure caused by heat treatment; quenching; hardening of milling cutters; relative speed of cooling of skin and interior; applications of theory to practice. Bibliography.

ENGINE-VALVE. Valve Steels, P. B. Henshaw. Royal Aeronautical Soc. Jl., vol. 31, no. 195, Mar., 1927, pp. 187-210 and (discussion) 211-217, 20 figs. Considers properties of valve steels in general use, and points out possible lines on which development is likely to take place; among properties which valve steel should possess are greatest possible strength at high temperatures; highest possible notched-bar values; forgeability; capacity of being manufactured free from cracks; and of being easily heat treated; least possible tendency to scale; ability to retain its original physical properties after frequent heatings to high temperatures followed by cooling to normal temperature; no liability to harden when cooled; capacity of being heat treated after forging; sufficient hardness to withstand excessive wear in the stem, etc.; author places steel for forging in following order: nickel-chromium, chromium, stainless, cobalt, high-speed and silicon-chromium steel.

FAILURES. Some Reasons Why Engineering Components Fail. Mech. World, vol. 81, nos. 2091 and 2092, Jan. 28, and Feb. 4, 1927, pp. 59-60 and 82-83, 6 figs. Failures can be divided into two distinct groups, those which are due to faulty designing, and those due to faulty material. Some of the more common and more important causes of failure.

FRACTURES. Interpretation of Steel Fractures, P. W. Priestley. Indus. Chem., vol. 3, no. 25, Feb., 1927, pp. 63-66, 5 figs. To insure correct interpretation of any fracture, it is necessary to study causes of defects and failures in metals; principal factors are chemical composition, treatment in casting stage, mechanical working, both hot and cold, subsequent heat treatment and abuse; occurrence and types of fractures; miscellaneous influences.

HEAT-RESISTING STAINLESS. Progress in the Development and Practical Application of Heat Resisting and Non-Corroding Steels, R. Hadfield. Inst. Mar. Engrs. Trans., vol. 39, Feb., 1927, pp. 1-47 and (discussion) 47-53, 16 figs. Describes different types of steels in their respective classes, giving idea as to progress already made with them in their application to industry; steel for steam turbines and steam fittings; non-corroding steels, with special reference to new type known as Era/CR steel, its chief advantage over chromium steel being more complete resistance which it offers to ordinary forms of corrosion due to air and water, and to great range of corrosive agencies to which it is resistant; this steel will display its qualities without special heat treatment, and without specially careful machining; gives list of chemical agents to which these steels are completely resistant.

LEAD COATING. Homogeneous Lead Coating for Steel. Iron Age, vol. 119, no. 14, Apr. 7, 1927, pp. 996 and 1044. Lead coating by process which eliminates use of soft solder

or tin, as flux to effect adhesion of lead to metal being coated, has been developed by Gross Lead Burning & Coating Corporation; this is being used commercially for applying coatings of lead to steel, brass and copper containers for holding corrosive acids.

OXYGEN CYLINDERS, FOR. Causes of Explosions of Steel Cylinders Containing Compressed Gas (Causes d'explosion des bouteilles en acier contenant un gaz comprimé), C. Fremont. Genie Civil, vol. 90, no. 10, Mar. 5, 1927, pp. 239-241, 6 figs. Account of explosion of steel bottles containing compressed oxygen; current tests reveal segregation and non-metallic inclusions; initial cause of such explosions is generally due to impurities in steel and embrittlement of metal; points out that in manfacture it is possible and easy, with aid of practical corrosion and embrittlement tests, to determine presence of defects and thus avert danger of explosion.

SCALING. Influence of Atmosphere and Temperature Upon the Scaling of Steel. Chem. & Industry, vol. 46, no. 8, Feb. 25, 1927, pp. 61T-72T, 11 figs. Contains following articles: Scaling By Air, Water Vapour, and Carbon Dioxide, C. B. Marson and J. W. Cobb. Prevention of Scaling by Hydrogen and Carbon Monoxide, H. T. Angus and J. W. Cobb.

STAINLESS. The Machining of Stainless Steels. Machy. (Lond.), vol. 29, no. 754, Mar. 24, 1927, pp. 815-817. Turning and screwing stainless steel; stainless steel on turret lathes.

STANDARDIZATION. The Standardization of Steel (Stahlnormung), F. P. Fischer. Maschinenbau, vol. 6, no. 1. Jan. 6, 1927, pp. 5-10. 5 figs. Emphasizes importance and necessity of standardization; review of present stage of development; comparison of American standards for materials with German standards shows that former are much more comprehensive and in some instances give much more accurate data on composition and heat treatment of steel; points out that satisfactory standardization cannot be effected by steel manufacturers alone, but must be guided as well by many-sided demands of users.

STAYBOLT. Metallurgy of Staybolt Iron and Steel, F. Williams. Boiler Maker, vol. 27, no. 3, Mar., 1927, pp. 73-74, 12 figs. Heat treatment and physical properties of steels that are desirable for locomotive use.

TORSIONAL DEFORMATION. Phenomena Occurring in Twisted Steel Bars When Flow Limit has been Exhausted (Die Vorgänge nach der Ueberschreitung der Fliessgrenze in verdrehten Eisenstäben), W. Bader and A. Nadia. V. D. I. Zeit., vol. 71, no. 10, Mar. 5, 1927, pp. 317-323, 53 figs. Rules for formation of flow strata in twisted iron bars; position of thin flow lines in etched cross sections of twisted bars; spreading of flow strata in bars with different cross sections is experimentally determined.

STEEL CASTINGS

PROPERTIES. Cast Steel (L'Acier Moule), E. Oberg and Monfray. Fonderie Moderne, vol. 21, Mar., 1927, pp. 7-8. Annealing of castings; composition of molding steels; manganese steel.

SOUNDNESS. Section Affects Soundness of Steel. Forging-Stamping-Heat Treating, vol. 13, no. 3, Mar., 1927, pp. 101-102. Tests from forgings and castings show varying characteristics dependent upon location and section of pieces for tensile or chemical determinations.

STEEL, HEAT TREATMENT OF

MOLYBDENUM STEEL. Molybdenum Steels Heat Treatment. Soc. Automotive Engrs. Jl., vol. 20, no. 1, Jan. 1927, pp. 14-15. S. A. E. steels 4130 and 4140 covered by Iron and Steel Division Report.

OVERHEATING. Study of Overheating of Steels (Etude sur la surchauffe des aciers), L. Aisenstein. Revue Universelle des Mines, vol. 13, no. 3, Feb. 1, 1927, pp. 106-115, 10 figs. Conclusion of work published in same journal in 1925 (vol. 8, no. 6, Dec. 15, 1925); present study deals with nickel and chrome steels.

OXYACETYLENE FLAME. Heat Treatment with the Torch, E. E. Thum. Acetylene Jl., vol. 28, no. 10, Apr., 1927, pp. 473-476 and 478, 7 figs. Hardening of small tools; relationships between metal and flame; hardening; drawing or tempering; hardening of malleable iron; annealing.

TYPEWRITER PARTS. How Typewriter Parts Are Heat Treated for Long Life, R. L. Manier. Iron Trade Rev., vol. 80, no. 12, Mar. 24, 1927, pp. 769-770, 3 figs. Describes heat treating and japanning installation at L. C. Smith and Corona Typewriters Inc., Syracuse, N. Y.; gas is employed as fuel and has been found most satisfactory; furnaces of double-chamber type are used in tool department.

STEEL MANUFACTURE

BASIC. The Blast Conditions with the Basic Process (Die Windverhältnisse beim Thomasverfahren), H. Bansen and B. Von Sothen. Stahl u. Eisen, vol. 47, no. 10, Mar. 10, 1927, pp. 385-389, 5 figs. Determination of pipe, base and bath resistance and their respective parts in total pressure loss; influence of elevation of base on pressure loss, rate of blast capacity delivered per minute, and duration of blast.

OPEN-HEARTH PROCESS. Steel, F. G. Martin. Liverpool Eng. Soc. Trans., vol. 47, 1926, pp. 143-170, 25 figs. Deals with manufacture of Siemens-Martin steel.

STRUCTURAL STEEL

HEAT TREATMENT OF. The Heat Treatment of Structural Steel (Ueber die Wärmebehandlung von Konstruktionsstählen), F. W. Duesing. V. D. I. Zeit., vol. 71, no. 9, Feb. 26, 1927, pp. 297-300, 15 figs. Results of extended investigations carried out at Kaiser Wilhelm Institut für Eisenforschung.

SILICON. Silicon Steel Produced in the Open-Hearth Furnace (Silizium-Baustahl aus dem Siemens-Martin-Ofen), J. Meiser. Stahl u. Eisen, vol. 47, no. 11, Mar. 17, 1927, pp. 446-48. Results of strength testing of silicon steel show that material produced in open-hearth furnace of usual design, meets all requirements.

WELDING VS. CASTING. Replacing Castings with Structural Shapes, W. L. Warner. Am. Welding Soc. Jl., vol. 6, no. 2, Feb., 1927, pp. 36-44, 11 figs. Discusses applications of welded structural steel parts to replacement of castings.

TESTING MACHINES

CALIBRATION. Calibrating Instruments for Strength-Testing Machines (Eichgeräte für Festigkeitsprüfmaschinen), W. Wilk. Stahl u. Eisen, vol. 47, no. 10, Mar. 10, 1927, pp. 409-410. Deals with different types of equipment for calibration of materials-testing machines.

THERMOMETERS

MELTING-POINT MEASUREMENT. Apparatus for Melting Point Determinations (Een eenvoudig toestel voor smel tpuntbepaling), H. Ter Meulen. Chemisch Weekblad, vol. 24, no. 3, Jan. 15, 1927, p. 36, 1 fig. Thermometer with capillary tube attached is supported in long test-tube, suspended in wider tube with constriction near lower end; small burner is placed below constriction and piece of gauze above it.

THERMIT WELDING

LARGE PIECES. Thermit Welding, J. H. Deppeler. Tech Eng. News, vol. 8, no. 2, Mar. 1927, pp. 59 and 100, 2 figs. Welding of large pieces by this method.

TIN PLATE

TIN-IRON ALLOY IN. The Tin-Iron Alloy in Tin Plate, E. F. Kohman and N. H. Sanborn. Indus. & Eng. Chem., vol. 19, no 4, Apr. 1927, pp. 514-518, 6 figs. There is no consistent variation evident in service value in comparing coke cans and various lots of cans with more tin alloy, while charcoal cans show appreciably better service; all lots of tin cans, even charcoal cans, showed much greater tendency to perforate than untinned cans, while latter showed much greater tendency to hydrogen formation; reason for this, as well as better service value of charcoal cans, is discussed.

TUBES

STEEL, SEAMLESS. Influence of Coal Working on Strength Properties and Structure of Seamless Steel Tubes (Einfluss des Kaltziehens auf die Festigkeitseigenschaften und das Gefüge von nahtlosen Stahlrohren verschiedener Vorbehandlung), A. Pomp and W. Albert. Stahl u. Eisen, vol. 47, no. 11, Mar. 17, 1927, pp. 459-463, 8 figs. Results of tests on seamless tubes which have undergone different preliminary treatments; contains supplement on power required in tube drawing. Abstract of paper read before Kaiser-Wilhelm Institut für Eisenforschung.

WELDING

CENTRAL STATIONS. Welding Reduces Power House Costs, T. E. dePew. Welding Engr., vol. 12, no. 3, Mar., 1927, pp. 25-27, 8 figs. Lower building, operating and maintenance charges are insured by its use at N. Y. Edison East River station.

JOINTS. Some Facts About Welded Joints, S. J. Melville. Welding Engr., vol. 12, no. 3, Mar. 1927, pp. 37-38. Old superstitions about greater reliability of rivets are gradually giving way before new evidence in favor of welds.

MACHINE CONSTRUCTION BY. Building Special Machines by Welding, R. E. Kinkead. Am. Mach., vol. 66, no. 10, Mar. 10, 1927, pp. 409-411, 3 figs. Single casting requirements for special machines run up unit cost; substituting built-up steel parts; how welding is applied to special-machine construction.

PIPE. Pipe Welding Under Difficulties, G. Walker. Welding Engr., vol. 12, no. 3, Mar. 1927, pp. 29-30, 9 figs. Describes high-pressure steam pipe line which was welded under extremely severe winter weather conditions; welded joint chosen as best for high-pressure line; pipes connected outdoors in midwinter; line tests free of leaks.

RODS. Effect of Surface Materials on Steel Welding Rods. Boiler Maker, vol. 27, nos. 2 and 3, Feb. and Mar., 1927, pp. 51-52 and 77-78. Feb.: Discussion of effect on working qualities of metallic and nonmetallic surface materials; gas filler rods. Mar.: Study of various welding-rod surfaces, their characteristics and properties for certain welding operations.

RODS. Materials for Welding Rods (Werkstoffe für Schweisstäbe), Kantner. V. D. I. Zeit., vol. 71, no. 8, Feb. 19, 1927, pp. 253-254, Investigation of influences of rod covering on quality and economy of weld; expense and performance conditions; research work carried out by Committee on Welding of V. D. I.

STEEL CASTINGS. The Selection of a Welding Process, L. E. Everett. Forging-Stamping-Heat Treating, vol. 13, no. 3, Mar., 1927, pp. 98-100. Reasons for welding steel casting are considered; its use is not extensive, but for small defects is justified.

STEEL PLATE. Different Methods of Welding Steel Plates (Les différentes méthodes d'exècution des soudures sur toles d'acier). Revue de la Soudure x Autogène, vol. 19, no. 156, Feb., 1927, pp. 1301-1307, 28 figs. Describes different methods and gives examples of their application.

STELLITING. Stelliting: A New Welding Process, A. V. Harris. West. Machy. World, vol. 18, no. 3, Mar., 1927, pp. 104-105 and 137, 6 figs. Method of surfacing metal parts with stellite which effects important savings wherever utilization of remarkable properties of alloy is productive of economy.

WIRE

ANNEALING AND PICKLING. Annealing and Pickling of Iron and Steel Wires (Das Glühen und Entzundern von Eisen und Stahldrähten), P. Krumme. Maschinen-Konstrukteur, vol. 60, no. 3, Feb. 15, 1927, pp. 66-68, 5 figs. Discusses different annealing process and compares types of furnaces employed; cleansing of wires after annealing, usually by pickling with sulphuric or other acids.

WIRE DRAWING

STEEL. Relation of Mechanical Properties of Drawn Steel Wire to Natural Hardness and Stretching (Die Abhängigkeit der mechanischen Eigenschaften gezogenen Stahldrahtes von der Naturhärte und der Reckbehandlung durch das Ziehen), W. Püngel. Stahl u. Eisen, vol. 47, no. 5, Feb. 3, 1927, pp. 172-182, 14 figs. Influence of carbon on drawing capacity and strength properties of wire; influence of method of drawing; testing of finished wires; influence of cold drawing; numerical method of determining production processes; drawing capacity increases with decreasing carbon content; tensile strength increases in hard as well as in drawn wire proportional to carbon content; results of these investigations show that former method of testing wire rope is not adequate.

X-RAYS

AUTOMOTIVE MATERIALS, EXAMINATION. Applications of X-Rays in the Automotive Industry, G. L. Clark, R. H. Aborn and E. W. Brugmann. Soc. Automotive Engrs. Jl., vol. 20, no. 2, Feb., 1927, pp. 291-304, 22 figs. Résumé of X-ray science and experimental results regarding practical applications to examination of automotive materials; nature and relationship of X-rays to light, properties of X-rays which can be utilized practically, and in production, including X-ray tubes; detection of imperfections in castings, development of improved casting technique by X-rays, substitution of castings for forgings and X-ray examination of rubber tires, valves and the like; ultimate crystalline structure of pure metals and alloys used in automotive industry, such as brasses, bronzes, steels, alloy high-speed steels, aluminous alloys, and the like; effects of lubrication as determined by X-ray diffraction; steel suitable for electrical purposes.

COPPER. Developments in the Manufacture of Copper Wire, J. R. Shea and S. McMullan. Am. Inst. Elec. Engrs. Jl., vol. 46, no. 4, Apr. 1927, pp. 346-355, 26 figs. Developments in manufacture of copper wire and describes copper rod and wire mill designed to meet new requirements; survey of copper-rolling and wire-drawing art; outline of work done by Bell System engineers in developing and designing new types of wire-drawing machinery.

TECHNICAL USES. Some Technical Uses of X-Rays, A. St. John. Indus. & Eng. Chem., vol. 19, no. 3, Mar., 1927, pp. 339-342, 4 figs. Nature and origin of X-rays, ways they can be used for technical purposes, and jobs which have been handled successfully; advantages and limitations of fluoroscopic and of photographic inspection.

HARDNESS TESTING OF STEEL BALLS BY MAGNETIC METHODS

(Continued from Page 690)

In conclusion it is a pleasure to acknowledge the courtesy of the Wilson-Maeulen Company of New York City in furnishing one of their Model 2A Rockwell hardness testers for use in this work. Appreciation is expressed for the kind co-operation of the New Departure Manufacturing Company of Bristol, Conn. and the Norma-Hoffman Bearings Corporation of Stamford, Conn. Thanks are also due Mr. Fred Bedford, a graduate student in the department, for assisting in taking some of the readings.

DISCUSSION

CHAIRMAN ROCKWELL: I should like to ask Dr. Williams, if when putting the ball between the magnet poles, that an artificial pole is created, that is, a certain north and south pole line. Can that be changed to any position without putting it through the solenoid to demagnetize it?

DR. S. R. WILLIAMS: That is, after the ball is once placed between the poles, can it be rotated without affecting the reading? No, then one obtains any reading. Put the steel ball in there once and pull it out and it will be magnetized, as shown with those filings. Suppose you turn it through 90 degrees and set it back again and then pull it out, you will have four poles. You can have a regular fuzzy-wuzzy all over it, as you saw with those filings.

CHAIRMAN ROCKWELL: I am thinking of Fig. 6 showing the striations of the fracture. Have you broken any that had been subjected to that movement in different positions to note the type of fracture in reference to striations.

DR. S. R. WILLIAMS: No, I have not; that would be an interesting point.

News of the Society

UNAUDITED PROFIT AND LOSS STATEMENT AMERICAN SOCIETY *for* STEEL TREATING

For the period from January 1 to March 31, 1927

INCOME			
Gross Dues		$ 15,305.58	
Transactions Advertising	$10,529.32		
Transactions Sales	339.80		
Transactions Subscriptions	649.51	11,518.63	
Reprints		450.30	
Bindery Account		467.50	
Book Account		88.74	
Miscellaneous Receipts		129.01	
Data Sheets	696.40		
Data Sheet Binders	328.70	1,025.10	
Discounts Received	117.39		
Interest	57.86	175.25	
Sustaining Exhibit Membership		3,875.00	
TOTAL INCOME			$ 33,035.11
EXPENSE			
Local Chapters		$ 6,498.00	
Sectional Meeting—Washington		695.97	
Transactions		7,230.90	
Reprints		903.30	
Bindery Account		422.80	
Index		400.00	
Book Account Expense		72.51	
Library		27.75	
Data Sheets		1,134.12	
Data Sheet Binders		5.70	
Discounts Allowed		143.03	
Bank Exchange		.55	
E. D. Campbell Memorial Lecture		2.30	
General Expense		1,617.72	
1926 Convention Expense		12.25	
Insurance		253.75	
Secretary's Office		4,119.72	
President's Office		51.88	
Treasurer's Office		781.15	
Directors' Expense		1,141.82	
National Committees		1,602.89	
TOTAL EXPENSE			$ 27,118.11
EXCESS INCOME OVER EXPENSE			$ 5,917.00

American Society for Steel Treating

BALANCE SHEET

As of March 31, 1927

ASSETS			
Commercial Account—Cleveland Trust Co.		$ 13,898.82	
Savings Accounts		28,925.58	
Accounts Receivable		9,837.25	
Accrued Interest		297.73	
Investments		67,846.25	
Office Furniture and Fixtures		1,267.59	
Inventory		4,129.95	
1927 Convention Prepaid Expenses		4,746.11	
TOTAL ASSETS			$130,949.28
LIABILITIES, RESERVES AND SURPLUS			
Accounts Payable		$ 2,572.01	
1927 Convention Advance Receipts		9,203.29	
Reserves		35,515.05	
Surplus—January 1, 1927	$77,538.05		
Plus—Credit Balance from Surplus Adjustment	203.00		
—Excess Income over Expense for the period from January 1 to March 31, 1927, per profit and loss statement attached	5,917.00	83,658.93	
TOTAL LIABILITIES, RESERVES AND SURPLUS.			$130,949.28

News of the Chapters

STANDING OF THE CHAPTERS

DURING the month of March there were 160 new and reinstated members added to the rolls of the society. There were 45 old members dropped for non-payment of dues and 27 for resignations and deaths, leaving a net gain for the month of 88 compared with a net gain of 93 during March, 1926.

In the following tabulation there appears the relative membership standing of the 30 chapters and 2 groups of the Society as of April 1, 1927.

GROUP I		GROUP II		GROUP III	
1. Detroit	421	1. Hartford	138	1. New Haven	76
2. Chicago	402	2. Milwaukee	131	1. Tri City	76
3. Cleveland	400	3. Lehigh Valley	127	3. Los Angeles	75
4. Philadelphia	351	4. Golden Gate	117	4. Washington	70
5. Pittsburgh	337	5. Indianapolis	97	5. Rockford	65
6. New York	277	6. Montreal	94	6. Rochester	60
7. Boston	254	7. Cincinnati	93	6. Toronto	60
		8. Syracuse	88	6. Worcester	60
		9. St. Louis	87	9. Providence	54
		10. Buffalo	64	10. Southern Tier	52
		11. Dayton	55	11. Schenectady	46
		12. North-West	54	12. Fort Wayne	43
				13. Springfield	32

GROUP I—Detroit still heads this group with a total of 421. Chicago advanced to 2nd place with a net gain of 7 members. Cleveland drops to 3rd place, although they had a net gain last month of 10 new members. However, they lost 11 members by transfer from Cleveland to Dayton. It is interesting to note that, although the Cincinnati and Cleveland chapters lost a total of 17 members to Dayton, both chapters sent the new organization their heartiest congratulations and best wishes. Philadelphia had a loss of 3 members. Pittsburgh had a gain of 7; New York 5; and Boston 8. The standing of the chapters, however, with the exception of Chicago and Cleveland, remains the same.

GROUP II—At their own request, Dayton was placed in Group 2, and Montreal, also at their own request, was transferred from Group 3 to Group 2, and, with their 94 members, assumes position No. 6. Hartford, although suffering a loss of members last month, still heads the group. Milwaukee, with a net gain of 2, assumes 2nd place, displacing Lehigh Valley, which had a net gain of 4. Golden Gate and Indianapolis still have positions 4 and 5, while Montreal goes in as No. 6. Cincinnati, Syracuse, St. Louis and Buffalo still have the same positions, while Dayton goes in as No. 11 with 55 paid-up members.

GROUP III—By the transfer of Montreal, the contest for position No. 1 is with New Haven, Tri City and Los Angeles, with Washington a close contender. Los Angeles, which was No. 2 in the last report, has now dropped to No. 3, while New Haven with a gain of 2 and Tri City with a gain of 4 go into a tie for first place. We received a wire from Tri City stating that 11 new applications were on the way, but these could not be counted in the March report because remittances were not received. But, with the billing of these 11 new members for April, it behooves New Haven and Los Angeles to watch their laurels, or this interesting contest for the top of the group will be decided for a time at least in favor of Tri City. Rochester, Toronto and Worcester have developed into another "triple tie", all three chapters having 60 members each. Southern Tier showed a good gain last month of 6 new members and passed Schenectady. Fort Wayne gained 3 new members, while Schenectady lost 4 and Springfield 1. The total membership of the Society on April 1 was 4524.

BOSTON CHAPTER

The April meeting of the Boston Chapter was held on Friday, April 1, 1927, the afternoon was spent at the Mystic Iron Works in Everett and the evening meeting was held at Massachusetts Institute of Technology.

Through the courtesy of Thomas Kennedy, president and general manager of the Mystic Iron Works, the chapter was invited to witness the tapping of their blast furnace and see the other operations around the plant. This development, which is causing much interest in New England, represents an investment of $5,500,000, and is designed for an annual capacity of 135,000 tons of machine-cast pig iron for foundry use. Ore for the plant is being imported from Africa, Newfoundland and Sweden, while the coke comes from the nearby plant of the New England Coal and Coke Company.

Promptly at 1:30, the furnace was tapped and a crowd of about 125 foundrymen and heat treaters watched the metal being run into two fifty-ton ladles and afterwards into the pig casting machine. A very elaborate power plant is maintained to generate steam for the turbine blowers, small coke and breeze being used for fuel. A completely equipped chemical laboratory analyzes all of the incoming and outgoing materials.

Dinner was served to seventy members and guests in the Walker Memorial at 6:30 p. m. A short business meeting was held, during which the names of the officers for the coming year, nominated by the executive committee, were read. The annual election will be on May 6, at which time additional nominations will be received from the floor.

The speaker of the evening was Dr. Enrique Touceda, consulting engineer for the American Malleable Castings Association at Albany, New York. Dr. Touceda selected as his subject "The Manufacture of Malleable Iron Castings and Details Regarding the Product," and discussed the selection of raw materials, melting and molding practice, annealing, testing and inspection. A complete history of the evolution of melting furnaces was given and the characteristics of the modern air furnace fully explained. Many slides were shown demonstrating the effect of design and gating on the shrinkage of the casting. Shrinkage is the malleable maker's worst enemy, partly due to the loss from

breakage and also due to the fact that the cavities retain their hardness after annealing and ruin the machine tools.

Charts were shown demonstrating the great increase in physical properties which have been obtained since the industry introduced organized research. The speaker also stated that it has been conclusively demonstrated that the strength of a malleable casting is not in the outside skin as sometimes claimed, as just as good physical properties are obtained if this outside surface is removed before testing.

The visitation and meeting were under the general direction of H. B. Parker, chairman of the program committee, and the meeting was presided over by vice-chairman L. D. Hawkridge. On the recommendation of the executive committee, a letter has been written the president of the Massachusetts Institute of Technology, thanking him for his generosity in again placing the facilities of the Institute at the disposal of the chapter during the past year. Thanks are also due to Professors Waterhouse and Williams for their very valuable assistance to the program and meetings committee.

H. E. Handy.

BUFFALO CHAPTER

The meeting was held on March 31, 1927, and called to order by Chairman Armstrong at 8:30 p. m. The minutes of the meeting of March 4, 1927, were read and approved.

Captain Godfrey of the Buffalo Traffic Department gave a very interesting talk on "Street Signals." An interesting discussion followed.

R. H. Smith, of Lamson and Sessions Company, was introduced to discuss "High Tensile on Low Carbon Steels by Heat Treating." Mr. Smith explained the methods of heat treating and the results which were obtainable. Lantern slides were in use to show machines, test specimens, and results. Much discussion followed.

About fifty members and guests were present.

Lunch was served after adjournment.

B. Clements.

DAYTON CHAPTER

The April meeting of the newly organized Dayton Chapter was held Monday evening, April 18, 1927, in the auditorium of the Dayton Engineer's Club. About 100 members and friends heard Mr. W. S. Bidle's interesting and practical talk on "The Steel Treater's Tools." Before starting his subject, Mr. Bidle spoke entertainingly of the early history of the Society and of the difficulties encountered and overcome during the first year or two of its life.

As president of the W. S. Bidle Company of Cleveland, Mr. Bidle has traveled extensively and visited several hundred of the country's largest and most progressive steel working and treating plants including practically all of the largest automobile manufacturers and tool makers.

Many of the novel ideas in furnace design and operation were described to the audience and were illustrated by crayon sketches. In his prefatory remarks Mr. Bidle brought out the advantages of plant visitation in keeping the older and more practical man up-to-date and progressive.

The talk dealt mainly with furnaces and furnace design. The relative merits of oil, gas and electricity were discussed impartially, especially from the economic viewpoint. Mr. Bidle emphasized that the final choice of any of the three heating mediums depended on conditions in the particular locality. Attention was called to the necessity of storage and of handling charges for oil and to the decreased life of refactories in oil fired furnaces.

Furnaces designed for both oil and gas were described in detail; with special attention to the spacing and arrangement of the burners. These furnaces are installed extensively in localities where the gas supply is apt to be scarce in winter. This temporary scarcity of gas in winter has, in the last few years, been almost completely overcome.

Mr. Bidle gave the electric heat treating furnace most of the credit for increased efficiency in furnace design. Due to the high cost of power electric furnace builders were, from the first, compelled to pay special attention to insulation. This resulted in calling the attention of the gas and oil-fired furnace builders to the importance of this factor.

During the course of his talk Mr. Bidle described and illustrated under-fired and over-fired furnaces, rotary carburizing furnaces, and discussed increased efficiency in heat treatment through application of the counter flow principle.

Several novel ideas in cooling quenching oil were given to the audience. A lively discussion centering mainly around rotary carburizing furnaces closed the meeting.

The next meeting, May 16, will be addressed by Earl C. Smith, assistant general superintendent of The Central Alloy Steel Corporation. Members of other chapters visiting in Dayton are cordially invited to attend.

F. T. Sisco.

INDIANAPOLIS CHAPTER

At the regular monthly meeting of the local chapter of the Society we were pleased to have as our guest S. P. Rockwell who delivered a most interesting address on the Rockwell Dilatometer and the Volcrit method of heat treatment. Because of the very nature of the subject, we had one of the very best meetings of the year and many of our members and their guests from nearby cities and towns were present. Mr. Rockwell illustrated in detail with moving pictures the way in which the Dilatometer operated and all present were able to see the numerous places of importance where this instrument could be used in order to assure the manufacturer the very best possible results to be obtained from the steel. Naturally his talk started considerable discussion in which many of the members participated and it is certain that all present were able to realize the necessity of controlling accurately the temperatures, if the highest quality products were desired.

The customary dinner was served at 6:30 p. m. after which many of those present were called upon to recite some of their experiences and several stories brought forth much laughter. A quartet composed of men affiliated with the Link Belt Company gave several selections. Unfortunately, Judge S. J. Carter, this city, was compelled to cancel his engagement for the evening because of illness.

Judging from the extreme interest manifested at the meetings and the number of new members being installed, it is evident that our local chapter is destined to be one of the influential scientific Societies in this district as well as imparting valuable knowledge to those who attend.

C. H. Beach.

LEHIGH VALLEY CHAPTER

The regular March meeting of the Lehigh Valley Chapter was held at the Exhibition Buildings of the Bethlehem Steel Company, Bethlehem, Pa., on Friday, March 18, at 7:45 P. M.

The speaker, J. A. Doyle, vice-president of the W. S. Rockwell Company, was our guest at the executive committee dinner held at the Bethlehem Club and later addressed our meeting.

The meeting was opened with an innovation in the form of a moving picture. We secured from the Bureau of Mines their film entitled "The Story of Power." If all of the films put out by the Bureau of Mines are as interesting as this one, we would suggest this as a very attractive supplementary feature for chapter meetings. Information regarding these films may be had by addressing the Bureau of Mines Experimental Station, Pittsburgh, Pa.

Mr. Doyle talked to us on the subject of heat treating equipment. It takes an artist to paint absorbing interest into an otherwise drab scene, and it apparently takes Mr. Doyle to create an epic out of a subject like heat treating equipment. This was one of the most interesting talks that we have ever been privileged to listen to. It had to do with fundamental principles of heating steel, the importance of which was secondary only to the interesting manner in which they were presented.

F. R. Palmer.

NEW HAVEN CHAPTER

Bridgeport Day of the New Haven Chapter on Thursday, April 7, has come and gone, but to many it was something more than just a meeting of the American Society for Steel Treating.

At 4 o'clock we assembled at the American Tube and Stamping Company of the Stanley Works on Stratford Avenue, and under the supervision of R. L. Baldwin we were escorted through the many departments to see just how steel was manufactured. To many the fabrication of steel is very interesting, and from the number of questions asked, we have no doubt the plant visitation was indeed a success.

Shortly after six o'clock, we motored to the University Club where a very tempting and tasty menu had been prepared by the committee. More power to committees of this caliber.

Just before eight o'clock we journeyed to the assembly room of the Bridgeport Brass Company, and after a short business meeting the chairman introduced the speaker of the evening, E. B. Estabrook, district manager, pyrometer division, of the Leeds and Northrup Company, Philadelphia, who gave a very interesting talk on "Temperature Measurement." This talk was illustrated, and from the number of questions asked after the meeting, there is no doubt but that the subject was well selected.

The attendance prize donated by F. E. Stockwell of the Standard Oil Company of New York was won by G. Gunther, of the Bridgeport Brass Company, Bridgeport. In addition to the regular attendance prize, a special one donated by Walter Underhill of the Regal Silver Company, New Haven, was won by John Bell of the Sun Oil Company, Bridgeport. New Haven is becoming known by these attendance prizes, and anyone attending the meetings is eligible for the prize, so you see it pays to be among those present.

The attendance was again good and the writer notices the R. Wallace & Sons Manufacturing Company, of Wallingford, and the Heppenstall Forge Company, of Bridgeport, were tied with six men each out to the meeting. The New Haven Clock Company, of New Haven, had four men present. Keep the good work up. More power to some of the other companies to try to beat these records.

The writer notices from the last monthly report from Cleveland that the New Haven Chapter is tied with Tri City for first place in Group 3, but the latter chapter has forwarded eleven new applications which will put them in first place. Come on, boys; let's get busy, and keep New Haven at the head of the list. Everybody try and bring in a new member. *W. G. Aurand.*

PITTSBURGH CHAPTER

The Pittsburgh Chapter held its annual "Round Table" meeting on the evening of April 7, in the U. S. Bureau of Mines Building.

After an executive committee meeting, the committee together with about forty members enjoyed an hour or so at a dinner which was served in the Bureau of Mines Restaurant. Many expressions of pleasure derived through the good-fellowship enjoyed at these dinners have been made.

At eight o'clock, Chairman W. H. Phillips called the meeting to order in the auditorium, and, after the usual short business session, announced the program as follows:

"Spark Testing" by W. B. Crowe.
"Spheroidizing" by Dr. C. M. Johnson.
"Brinell Hardness" by A. M. Cox.
"Carburizing" by W. J. Merten.
"Distortion in Hardening" by J. P. Gill.

Mr. Crowe, through the aid of a number of specimens and a high speed grinding wheel, demonstrated the characteristics of sparks of various carbon steels and said in part:

"There is no doubt but that everyone has witnessed a grinding operation accompanied with the usual display of sparks. Possibly there are some who have observed a radical difference in the character of the sparks without attaching any particular significance as to the cause of these differences.

"When steel is brought into contact with a rapidly rotating grinding wheel the grits of the wheel behave as minute cutting tools operating at extremely high cutting speed. When these particles of steel are hurled into the air with almost meteoric velocity they have been heated by mechanical action nearly to the molten state. Exothermic action then commences due to this initial heating to friction with the air and also the relatively large surface as com-

pared with mass, resulting in the combustion of the particle partially or completely.

"Obviously particles containing different percentages of carbon will behave differently when undergoing such an extremely rapid reaction. The products of combustion of the contained carbon bursts the molten particle with explosive violence, so that by a proper correlation between carbon content and character of sparks, a practical means of approximating the carbon content of steels is furnished.

"The principal characteristics to observe in spark analysis are: color of sparks; character of main or carrier spark as regards length; shape and color; and character of the explosions known as forking or sub-forking. Persistence of vision is probably a large factor in the appearance of the sparks, so that the impression of the same spark on different eyes may be somewhat different, but inasmuch as the test is purely relative the efficacy of the test is not impaired.

"A long carrier spark of dark red color and few if any explosions or forks, indicates a low carbon content. Where the carrier spark is short and brilliant, with initial forking close to the wheel, and also a multitude of sub-forkings, high carbon content is denoted.

"If a group of standards are prepared with 5 to 10 points of carbon differences, experience will soon enable one to arrange these specimens in their proper carbon content order by means of the spark. Having become proficient, obviously one would be able to determine the approximate carbon content of an unknown.

"Varying degrees of hardness of the same steel has no influence on the characteristics of the spark, inasmuch as before the particle has left the wheel it has ceased to be hard—or rather the particles have attained the same degree of hardness.

"The fusion point of high tungsten steel is elevated to such an extent that the exothermic reaction has scarcely a chance to be effective, resulting in dark red sparks with only an occasional fork.

"The practical value of the spark test is not difficult to see, especially if one has had the sad experience of mixed lots of steel which have frequently not been detected until the material has failed in service.

"Also, a test which can be made in the yard, in many cases without moving the material and, under favorable conditions, requiring less than one-half minute per test, is certainly entitled to serious consideration."

Dr. Johnson defined "spheroidizing" as the bringing together of the cementite (Fe_3C) particles existing in steel into a loose aggregate which takes on a globular or nodular form. This change is accomplished by prolonged heating slightly below the temperature at which steel will harden if suddenly cooled. The presence of considerable amounts of nickel, manganese or copper lower the latter temperature, while chromium, silicon and vanadian steels require a higher temperature, as also does tungsten steel. It is interesting to consider that if manganese be present to the extent of 13 per cent, or nickel to the amount of 25 per cent, no critical point can be detected in cooling these steels from high temperatures to room temperature. The rate of lowering of

the critical point by these two elements would, therefore, be about 30 degrees Cent. for each one per cent of nickel present, and about 50 degrees Cent. for each per cent of manganese present. However, the usual range of commercial steels can be spheroidized somewhere between (600 degrees Cent. and 740 degrees Cent.) 1112 Fahr. and 1364 degrees Fahr. It is usually sufficient to hold the steel at the favorable temperature for from 5 to 12 hours to secure perfect spheroidization. The low carbons below 0.50 per cent can be brought to the globular state in about half the time required for the eutectoid, i. e., 0.85 to 0.90 per cent carbon steel. The latter steel is supposed to be the most difficult to spheroidize, more so than the higher carbon steels. Sorbitic steels spheroidize more readily than the pearlitic steels for the reason that the cementite is in the finest state of division in the sorbitic steel. Steel, prior to holding below the critical range, should first be normalized; that is, heated to a temperature at which carbides will completely dissolve in the iron forming the austenitic steel. This is called heating through the Ac_{cm} range, or the temperature at which all of the cementite is dissolved in iron. The steel is then cooled as fast as possible in the air.

Also, steels with a carbon content in excess of the eutectoid, i. e., above 0.90 per cent carbon, are more quickly spheroidized by reason of excess cementite being present offering more nuclei. In the balling up process, the cementite does not travel far. One does not find the rounded bodies distributed sparsely or evenly throughout the much larger area of ferrite in the spheroidized steels of low carbon content, such as 0.20 and 0.30 per cent carbon. On the contrary, they are all located in islands that were pearlite areas previous to the annealing. As evidence that the spheroids are loosely knit, one finds that even though the drillings from such steels by reason of extreme softness are thick and chunky, yet these chunks dissolve with extreme rapidly in dilute nitric acid. If these balls were dense, the chemical annealing test would show the case as such, but instead, the test reveals the carbide to be in the finest state of division as to the individual particles that make up the spheroids.

This test was originated about thirty years ago by E. S. Johnson. It is an infallible test for complete spheroidization, or what is the same thing—perfection of annealing of steel in carbon steels. It can be made in five minutes or less. It consists of dissolving 0.100 gram of drillings with 5 c. c. of 1.20 Sp. Gr. nitric acid, in a perfectly clean test tube 6 inches long by ¾ in. diameter without heating. The action is rapid and the carbide particles are quite black at first and of the finest state of division, and do not collect together but are so finely divided that even a steel with a carbon content of as much as 1.20 per cent does not reveal its carbon in large brown flakes as it does when not annealed. Instead, the carbide particles are so fine that one almost wonders where they are when one suddenly finds they are mostly collected in the bottom of the tube in a fine powder, having settled imperceptibly. The solution and settling in a well-spheroidized sample takes about sixty seconds. By sharp contrast, a steel that has had little or no annealing will yield drillings with the same drill, that require three minutes to get over the

effervescence of gas from the acid attack; and thirty minutes to settle partly to the bottom of the tube.

Why is steel spheroidized? To get it into the maximum degree of softness, which is highly desirable for most machining operations. When steel is perfectly spheroidized it is as soft as it can be gotten, as will be shown by the Brinell hardness. There is only one softer condition for steel of appreciable carbon content; and that is attained by prolonging the heating to such an extent as to cause the cementite to decompose into the last and most stable phases; namely, elemental carbon or graphite, and iron.

If the soaking period is too lengthy and the silicon is in excess of 0.15 and carbon over 1.10 per cent, the graphitic phase is attained thereby rendering the steel unfit for keen-edged tools.

The speaker believes that spheroidizing proceeds at the greatest speed when the normalized steel is held as close to the temperature at which pearlite dissolves in iron to form austenite as possible without actually attaining to that temperature. In other words, not quite attaining to solid solution. This point could easily be settled by experiment.

Why does cementite assume the nodular form? One can see why it should collect in a mass because it cannot exist in the ferrite in solid solution. But why its rounded shape? Howe called the process "a true divorcing of pearlite into ferrite and cementite." Why does it ball up? The following may explain the globular outlines: "Let us consider the fact that spheroidizing takes place at a temperature when expansion is interrupted and contraction sets in; and, further, that in cooling down, contraction ceases suddenly and expansion occurs for a time. The cementite being very hard and brittle cannot stand the strain, and the carbide fence is shattered into tiny fragments. The contraction being more or less equal in all directions, these fragments are gradually compressed into more or less rounded bodies. That these forms are probably loose agglomerates of extremely minute fragments is suggested by the finely divided particles shown in the chemical annealing test already stated and the extreme rapidity of solution of the spheroidized steel."

Mr. Cox then took up the subject of "Brinell Hardness" and its relation to other physical properties, saying that Dr. Brinell was one of the first to prepare a comparative table of Brinell hardness and tensile strength.

Kent shows a relation between the Brinell hardness and tensile strength which was determined from experiments made in Sweden on annealed steels. The relations are as follows:

$$\text{"M"} = \text{"C"} \times \text{"H"}$$

where "M" is the tensile strength
"C" is a constant
"H" is the Brinell hardness

The value of the constant is as follows:

Impressions made transverse to direction of rolling and "H" is below 175, then "C" will be 462. The same direction of rolling with "H" above 175, then "C" will be 490.

Impressions made in direction of rolling and "H" is below 175, then

"C" will be 504. The same direction of rolling with "H" above 175, then "C" will be 516.

R. R. Abbott (American Society for Testing Materials, Vol. XV, Part II, 1915) evolved the following equations from a large number of tests, where:

"M" = Maximum Strength in 1000 lbs. per sq. in.
"B" = Brinell Hardness Number
"S" = Scleroscope Hardness Number

Carbon Steels	Nickel Steels
"M" = 0.73B — 28	"M" = 0.71B — 32
"M" = 4. 4S — 28	"M" = 3. 5S — 6

Chromium-Nickel Steels	Chromium-Nickel Steels
0.50% Cr 1.5% Ni	1.00% Cr 3.5% Ni
"M" = 0.68B — 22	"M" = 0.71B — 33
"M" = 3. 7S — 1	"M" = 3. 7S — 3

Chromium-Vanadium Steels
"M" = 0.71B — 29
"M" = 4. 2S — 21

In February, 1919, T. D. Lynch compiled data on Brinell hardness in relation to the other physical properties. The following formulae were derived from 1371 tests:

Tensile Strength = 434.6B + 11060
Elastic Limit = 251.7B + 5642
Elongation = 45.6 — .0982B

In introducing the subject of "Carburizing," W. J. Merten said, "that a survey of the literature on carburizing reveals a curious state of affairs in that practically the entire subject is handled from a standpoint of attainments of 'Technical Ideals' rather than from a standpoint of 'Economical Practicability.'

"However, carburizing is an art, and the human element plays a considerable part in the problems. What is technically possible is not always possible of commercial attainment."

In describing the mechanism of carburizing, the speaker said that "when iron and carbon are brought into contact and heated above 900 degrees Cent. they combine forming a solid solution of carbon in iron. In actual practice, carburization is carried on in a gaseous mixture of CO and CO_2.

"Carburization or cementation is produced by the following reduction of CO gas by iron:

$$3Fe + CO = Fe_3C + O$$

"The carbide, Fe_3C, going into solution in gamma iron until an equilibrium between the degree of saturation of austenite and the gas mixture at the prevailing temperature and pressure is established. It should be pointed out that, the larger the percentage of CO gas in the gas mixture (temperature and pressure remaining constant), the higher the saturation point of gamma iron with carbide will be raised, until a maximum saturation is obtained.

"When the gas phase, the system CO and CO_2, and the solid phase, iron,

are still unstable after such saturation is obtained, carbide (Fe_3C) will continue to form but free carbide will be diffusing and deposit along the grain boundaries of the austenite.

"Raising the temperature but keeping the gas mixture constant with respect to per cent CO gas will increase cementation but concentration of carbon in the case will not be raised although depth of case increases more rapidly since the higher temperature favors diffusion of the carbide (Fe_3C)."

The important considerations involved in carburizing are:

Carburizer—Carbon and Oxygen carrying medium
Energizers—Na or Ba Salts, Carbonates and CN Salts
Spacing of material in box
Luting of box
Time element of soaking
Cooling in box
Heat Treatment.

Owing to the lateness of the hour, J. P. Gill confined his talk on, "Distortion in Hardening," to a brief enumeration of the three characteristics which tend toward distortion as:

1. Loss of strength with increase in temperature
2. Increase in volume with increase in temperature and vice versa
3. Volume change incident with structural change.

This tendency is emphasized in hardening by one or a combination of three causes which may be simply stated but not so simply avoided or overcome:

1. Non-uniform heating
2. Non-uniform cooling
3. Strains not removed before hardening.

While giving the above, Mr. Gill graphically described each type of distortion.

The discussion of these papers was of necessity cut quite short and the meeting adjourned with a rising vote of appreciation to the speakers.

Harry A. Neeb, Jr.

ST. LOUIS CHAPTER

The sixty-fifth regular monthly meeting of the St. Louis Chapter was held Friday evening, March 25, 1927, with 67 members and guests present.

After the usual dinner the meeting was called to order by the chairman, Mr. Swander, and the secretary announced J. T. Ryerson and Son, Inc., as a new sustaining member.

After a short business meeting, Dr. F. C. Langenberg, research department, Climax Molybdenum Company, New York, was introduced. The title of his lecture was "The Cold Working of Hollow Cylinders, Centrifugal Casting and Gun Tubes." The chapter spent a very interesting evening listening to Dr. Langenberg's discourse on various subjects of cold working of metals.

This lecture was too long to be incorporated in this report, but for the information of members of the St. Louis Chapter and the Society as a whole, Dr. Langenberg gave this paper before the Cleveland Convention, and the paper in full can be read in the TRANSACTIONS of October, 1925.

This was an opportunity to hear about a remarkable metallurgical development, told by a man expert enough to talk in simple understandable language. We are sure that all the members who did not avail themselves to hear Dr. Langenberg missed a treat.

There being no further business, Dr. Langenberg was given a rising vote of thanks and the meeting was adjourned until April 15, at which time A. H. d'Arcambal of the Pratt and Whitney Company will lecture on metallurgical control in a plant manufacturing machinery, small tools, gages and air craft motors.

There will be a joint meeting with the St. Louis Chapter of the American Society of Mechanical Engineers at this meeting, and for the May meeting we have Dr. J. A. Mathews of the Crucible Steel Company.

C. G. Werscheid.

The sixty-sixth regular monthly meeting of the American Society for Steel Treating, St. Louis Chapter, was held jointly with the American Society of Mechanical Engineers, Friday evening, April 15, 1927, at the American Annex, with 102 members and guests present. This was indeed a fine attendance, in spite of the inclement weather, and it took nerve to brave the floods in this section.

After the usual dinner the meeting was called to order by Mr. Swander, chairman, after which the minutes were approved and read. No new business was reported.

The following nominating committee was appointed: F. G. White, National Enameling and Stamping Works, Chairman; L. T. Clark, Leschen Wire Rope Company; and G. C. Hall, Atlas Tack Corporation; who are to nominate the officers for the coming year.

The speaker of the evening, A. H. d'Arcambal, chief metallurgist Pratt and Whitney Co., gave an excellent practical talk illustrated with slides, the subject being "Metallurgical Control in a Plant Manufacturing Machinery, Small Tools, Gages and Air Craft Motors." Mr. d'Arcambal brought out some wonderful points on the proper hardening and grinding of various small tools. One high light in the small tool item was ground high speed taps after hardening.

One of the high lights of the meeting was brought forth in an air craft motor called the "Wasp," air cooled, used by the United States Navy, 650 pound engine with a 450 horse-power, with speed of 180 miles per hour, which has an acceleration of 600 to 1900 revolutions per minute in 4/5 of a second —or in other words, has an acceleration of a speed of 100 miles in 6 seconds, with a take-off of 120 feet. This was indeed an entertaining meeting, and all who were unable to attend certainly missed a treat.

There being no further business, Mr. d'Arcambal was given a rising vote of thanks, and the meeting was adjourned until the next, which will be some time in May.

C. G. Werscheid.

TORONTO CHAPTER

At the March meeting of Toronto Chapter held in the Consumers Gas Company Building, March 25, reports of progress on the Steel and Power

Show to be held in Varsity Arena, Toronto, August 31, September 1 and 2, under the auspices of the Toronto and Montreal Chapters of the American Society for Steel Treating, and eight other Canadian organizations, were made by W. J. Blair, of the Canada Cycle and Motor Company; J. W. McBean, Central Technical School; T. H. Young, Jessop Steel Company, and W. B. Dilly, Production Materials Ltd. Space is being rapidly sold, forty firms having already made reservation. Interest in the project is increasing daily as details of the technical sessions are made known, and it is expected that W. H. Eisenman, national secretary, will attend to outline the advantages Canadian industry may obtain from the American Society for Steel Treating activities.

Molybdenum steels was the subject discussed by the principal speaker of the evening, Allan Kissock, metallurgical engineer, Climax Molybdenum Company, New York. While primarily a mining engineer, and chiefly interested in that phase of the production of the metal, Mr. Kissock has absorbed much valuable information, particularly as to the uses and the troubles experienced with alloy steels.

From a series of slides a clear idea of the processes of mining were obtained, as well as some interesting history of the metal. Early in 1914, in Tucson, Ariz., a Texas cowpuncher found a heavy, yellow mineral in considerable tonnage of tailings from an old gold mining operation. The mineral proved to be wulfenite, a lead molybdate.

An Arizona attorney discovered several European firms were willing to pay a good price for concentrates of the material, and a small operation as a result was started.

"I became interested in the enterprise about this time," continued Mr. Kissock, "more particularly in connection with the converting of the concentrates into ferromolybdenum. It seemed only logical to remove the more than fifty per cent lead and what little gold was contained, and deliver only a finished product. A smelter was erected in Tucson, of which both lead bullion and ferromolybdenum were the products.

"The process differed from regular lead melting only in that a soda slag was employed, and this operation was conducted in both a reverberatory and water jacketed blast furnace. The resulting soda molybdate slag was smelted with iron in electric furnaces to ferromolybdenum."

Mr. Kissock gave in the course of his talk an interesting description of the troubles experienced, both in producing and marketing the metal, and detailed its uses as an alloy as well as methods of treatment. A feature of the discussion was the frank way in which the speaker outlined the limitations within which questions must be kept if he was to pass along reliable information on subjects of interest to members, a stipulation which might well be incorporated in the "repertoire" of lectures in other fields.

Campbell Bradshaw.

TRI-CITY CHAPTER

The meeting was held at the Davenport Chamber of Commerce on Thursday evening, March 17, 1927. Thirty members of the chapter were present at dinner and later for a talk by R. S. Archer, metallurgist for the Aluminum

Company of America, Cleveland, Ohio. Mr. Archer's talk, which was illustrated with slides, pointed out the characteristics peculiar to the element aluminum and its alloys. By proper heat treatment of various alloys great strength and toughness may be secured in comparison with the properties of the pure metal. The alloys in most common use are those of copper and aluminum, silicon magnesium aluminum, etc. Mr. Archer explained in detail the chemical and physical changes these alloys undergo during heat treatment.

Mr. Archer also pointed out the future possibilities of aluminum alloys in the fabrication of machines requiring extreme lightness and now being made from wrought iron or steel. The advances made in both heavier and lighter-than-air aircraft have resulted largely from the use of these light yet strong aluminum alloys. The possibilities of these alloys are all the greater when it is considered that 10 per cent of the earth's crust contains minerals high in aluminum content.

WASHINGTON-BALTIMORE CHAPTER

At a special joint meeting with the Washington Chemical Society and the Washington Academy of Science on March 4, 1927, Prof. Cecil H. Desch of the University of Sheffield, favored the meeting with a most interesting and instructive talk on the subject "The Growth of Crystals." The lecture was illustrated with a number of slides.[1]

According to Professor Desch the geometrical side of the study of crystallography is at present almost perfect. It was, however, not until 100 years after Robert Hooke presented in his book "Micrographia" (1664) the first pictures of crystals as seen under the microscope, that the science of crystallography began to develop very rapidly. These phenomena of crystal growth still need much further study.

The modern conception of crystals is that they are composed of atoms arranged in a space lattice. The characteristics of the lattice determine the shape of the crystal and also its physical and mechanical properties. In a lattice, a unit of simple geometrical form is repeated at regular intervals in three dimensions. There are certain planes in which the atoms are more closely packed than in other planes. It is these planes of close packing which play the largest part in determining the growth and other properties of the crystal. It has been established that growth perpendicularly to a densely packed plane is always slower than growth in any other direction. A plane containing very few atoms is a surface of rapid growth. The faces perpendicular to directions of minimum velocity are those which survive. They increase in size at the expense of the others. Organic substances such as salol present good material for observation of crystal growth. Metals offer especial difficulties in this respect since they are opaque, and metallic crystals originate and grow only at elevated temperatures.

When a metal (copper) hemisphere provided with cooling arrangements is immersed in a fused mass and is kept at a temperature below the melting point of the fusion, the influence of undercooling and of the rate of removal of the latent heat of fusion may be studied. Crystal faces begin to appear as

[1]For details see "Growth of Metallic Crystals" by Cecil H. Desch, read at the February meeting of American Institute of Mining and Metallurgical Engineers. This paper is quoted here at length.

small flat areas on the curved surface. The plane faces have at first circular or oval outlines, but as they increase in size they meet along straight boundaries. Finally the hemisphere becomes a polyhedron, the faces of which are the stable faces of a single crystal of the substance.

The forces which hold the atoms together in a space lattice are in all probability identical with chemical forces, and depend on the electronic arrangements. Within the crystal these forces are symmetrical, but at a free surface they must have an unbalanced character, the result being the existence of "residual affinity" at the surface which therefore has the power of attracting and holding other atoms or molecules. The holding of similar atoms constitutes growth, but foreign atoms, ions, or molecules may also be held, thus modifying the character of the surface of the crystal. A layer of foreign molecules being held "absorbed" by certain faces, growth perpendicular to those faces is hindered, so that they become more persistent. For instance, urea forming complex molecules with sodium chloride is held more strongly by octahedral faces consisting wholly of sodium atoms or wholly of chlorine atoms than by the neutral cubic faces. Lead nitrate which crystallizes in colorless octahedra shows new colored blue faces when methylene is added.

The direction of chemical attack in a crystal, like its mechanical properties, is determined by the position of its most closely packed planes. The different reagents do not necessarily attack the crystal in the same directions, so that the distribution of chemical affinity on different planes has to be taken into account as well as the closeness of packing. The action of the reagent on a face is localized at points which develop into etching pits. These etch figures show the remarkable phenomenon of periodicity.

When the etched section of a steel ingot is carefully examined, there is seen to be a thin outer skin of extremely minute crystals, followed immediately by a zone of large columnar crystals, appr ximately at right angles to the surface. In ingots poured very hot ("scorched") this zone may extend to the center. The original nuclei may have any orientation, but only those which have a direction of rapid growth perpendicular, or nearly so, to the wall survive. In most ingots and castings the columnar zone does not reach to the center and the remainder of the mass solidifies in crystals which have a random orientation, so called "equiaxed" crystals. It seems most reasonable to suppose that the columnar crystals have separated under "metastable" and the equiaxed crystals under "labile" conditions.

It is believed that the production of nuclei is assisted by the existence of groups of molecules, already attached to one another, in the liquid state, above the melting point. It is suggested also that a high temperature breaks up the molecular aggregates which most readily form crystal nuclei. In the midst of the liquid, where equiaxed crystals are formed there is approximately equal spacing of the nuclei so that the size of the resulting crystals does not vary very widely about a mean value.

It seems that there must be a certain periodicity of recurrence of the conditions under which a nucleus can form. It has been previously shown by the speaker[2, 3] from a statistical comparison of the shapes of crystal grains in

[2] C. H. Desch. Second Report to Beilby Committee, Jnl. Inst. Metals (1919) 22, 241.
[3] C. H. Desch. Rec. Trav. Chim. Pays-Bas. (1923) 42, 822.

cast metals and those of foam cells, that the form of the boundaries is determined by the interfacial tension between the growing crystals and the liquid metal. For instance, isolated grains of brass, disintegrated by immersion in mercury, were examined, each face being marked and the number of its sides and the direction of its curvature being noted. A general resemblance was found between the shapes of soap foam cells and metal grains.

Minute gold crystals prepared by chemical precipitation lose their sharp outlines on heating at about 700 degrees Cent.—more than 300 degrees below the melting point—and become rounded and finally spherical. The surface tension and cohesion diminish as the temperature is raised, and it seems that the two properties do not decline at the same rate, so that the influence of surface extension becomes greater at high temperatures, when the cohesion is relatively weaker. It is probably the same cause which produces the rounding of dendritic crystals in alloys.

Passing to the question of the dendritic growth, the speaker pointed out that in non-metallic substances this is usually caused by viscosity of the liquid or by exceptionally rapid crystallization. It cannot be viscosity, however, which is responsible for the dendritic habit of metallic crystals. The relatively high heat of crystallization of metals is no doubt a factor. A most remarkable feature of the dendrites is their regularity. Under favorable conditions the branching takes place at regular intervals, although the perfection of snow flake crystals, which are dendrites with hexagonal symmetry, is never approached by metals.

It must be supposed that the crystal exhausts the material in its neighborhood and an appreciable time is needed for the conditions of growth to be reestablished; this recurring at regular intervals is one of many periodic phenomena which are encountered in chemistry. The formation of Liesegang rings in precipitates is the most familiar example.

During the discussion it was explained by Professor Desch that the peculiar etch pattern observed[5] on rolled and then annealed strips of pure tin upon corrosion in normal alcoholic stannous chloride and 0.64N hydrochloric acid, which was independent of and superimposed upon the crystal structure pattern, may be caused by convection currents.

At its March meeting, on the 25th, Dr. D. A. Lyon, chief metallurgist and supervisor of stations of the United States Bureau of Mines, addressed the Washington-Baltimore Chapter on the subject "Nation-Wide Activities of the Bureau of Mines."

In his introduction some statistical figures were given to remind the members of the importance, scope and influence of the mineral industry on modern civilization. Attention was called to the fact that each year about two billion tons of mineral substance (including oil) are taken from the earth, having a marketable value of nine billion dollars. Coal represents 61 per cent of the value of tonnage mined; petroleum, 12 per cent; iron, 6 per cent; copper, 3

[5]For details see Discussion of V. N. Krivobok and O. E. Romig's paper entitled "Surface Structure Versus Inner Structure of Metals" by A. I. Krynitsky. Trans. American Society for Steel Treating, Nos. 1924, p—630-633; and "Corrosion Patterns on Cold Worked Tin and Zinc" by Henry S. Rawdon, Alexander I. Krynitsky and Julius F. T. Berliner. Chemical and Metallurgical Engineering, 1922, p—212-213.

per cent; gold, 3 per cent. The United States contributed 39 per cent of the world's tonnage with an annual production of nine hundred million tons. It has within its boundaries more than one half of the smelting capacity of the world and 70 per cent of the oil refining capacity. The total value of the United States mineral products is five billion dollars annually, of which amounts the fuels represent three-fifths, the metallics, both ferrous and nonferrous, one-fifth, and the non-metallics, one-fifth.

This country imports minerals from all parts of the world, nearly equal in value to those which it produces.

Although definite information is difficult to obtain as to the number of men employed in mining, smelting or otherwise preparing mineral products for the market, it is safe to estimate that one-half per cent of the population of the United States is directly concerned.

Taking into consideration the fact that the use per capita of metals in this country doubles about every ten years and likewise the fact that probably not a half dozen great mines have been made in the last decade, it can be readily understood why one of the most important functions of the Bureau of Mines at the present time is to increase safety, health, economy and efficiency in the mining, quarrying, metallurgical, and miscellaneous mineral industries of the country.

The situation now confronting the mining and metallurgical industry is as follows:

The ores in this country are becoming lower and lower in metallic content, due to the fact that the richer ores were mined first. (Ore has been defined as metal-bearing rock that can be worked at a profit). It is for this reason that the Bureau of Mines, in its technological, economic and safety branches, is endeavoring to determine by research work and inquiries, to what extent the fundamental laws of physics, chemistry, and economics may be applied to mining, metallurgical treatment, and marketing of metals and non-metals, so that they can be mined, treated and sold at a profit. In this connection it is to be remembered that the same elementary laws of chemistry or economics apply to the petroleum as to the cement industry.

As previously indicated, the Bureau is divided into five major units, the technologic, economics, health and safety, and administrative branches and the office of the chief mining engineer. Each branch is composed of several divisions and these divisions are in turn sub-divided into sections.

The metallurgical division contains the following sections:—physical-chemical laboratory, lead and zinc, low grade complex ores, coal beneficiation, copper, ore beneficiation, experimental blast furnace, steel making and electric furnace, non-metallic, oxygen production, rare and precious metals.

The activities of the different stations were explained, and the results secured at those dealing with purely metallurgical problems were described in some detail. Among these were the study of the flotation process for ore beneficiation which has already revolutionized the recovery of values from many deposits of metallic minerals; the studies of reactions within the blast furnace and the most probable benefits to be derived from oxygen enrichment of the blast and the use of oval or rectangular rather than circular furnace

sections; the process developed for the manufacture of sponge iron from low grade ores and low grade solid fuels, and the work looking forward toward the commercial production of improved refractories.

During the discussion, the possibilities of replacing nickel-containing steels by molybdenum steels was considered and the abundance of molybdenum ores in the United States, brought out. The possibilities of producing alloy steels by direct reduction of mixed ores, were also discussed, particularly the problems involved when dealing with mixtures of compounds of metals of widely different ease of reduction such as iron, titanium, and molybdenum.

A. I. Krynitsky.

WORCESTER CHAPTER

The eighth regular meeting of the Worcester Chapter was held March 24 as a joint meeting with the American Society for Mechanical Engineers, Electrical Engineers, and Civil Engineers. Dinner was served at 6:30 p. m. to two hundred members and guests.

Professor D. S. Kimball, Dean of the College of Engineering at Cornell University, was the speaker for the evening. He is one of the outstanding engineers of the present day, and we were very fortunate to have him with us. The subject of his talk was "The Engineer and National Progress."

Dean Kimball pointed out that people progress by explaining their experiences either by the spoken word or by written records. Within the last 300 years men have complied vast records by keeping their experiences in written form. They have gathered more scientific knowledge during this period than ever before. With this knowledge they have built up modern civilization.

The lines of growth are three in number: *First,* Development of labor saving machinery. The remarkable development in the ways and means for producing power in the last fifty years is amazing, and we do not know what will happen in the next fifty years.

Second, Development of time-saving machinery. This includes automatic and semi-automatic machinery. How much further we can go in the development of this kind of machinery is unknown to us.

Third, Line of development. Transmission of intelligence. From signal to telegraph to telephone and then radio and radio is only a forerunner of other lines to come. All this has for us a high degree of civilization which no other country can boast. The engineer has accomplished this—of course, using the *word* engineer in a broad sense. Years ago it was an easy thing to define the work of an engineer, but today that is much changed as a great many men trained in engineering work find a great many lines of endeavor.

In all lines of development engineering ideas are being used. The diagram or graph that the engineer uses to express his reasoning has been applied to a great many lines of work. This graphic idea of expression has been developed within the last 25 years and has become an essential part of a high school pupil's training.

The field for the engineer is increasing and what is coming in the next fifty years in the form of development will depend a good deal on him.

G. Johnson.

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